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HAMILTON, ONTARIO, CANADA

DEPARTMENT OF METALLURGY AND MATERIALS SCIENCE

MICROCHEMISTRY AND TRANSFORMATION KINETICS OF EXPERIMENTAL BATCH ANNEALED DUAL PHASE STEELS

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INTRODUCTION

The submission of this report marks formally the end of the McMaster-CANMET contractual studies of batch annealed dual phase steels. The research has been fruitful and satisfying from the perspective of the university based researchers involved. As discussed later, the research has led to the development of new principles for the design of alloy compositions and processing routes for batch annealed steels; therefore, the implications for Canadian industry are tangible and attractive. The report is submitted in the hope that the technology will be tested next on a pilot plant scale and ultimately utilized in our steel industry, which now lacks the capability to produce these alloys without first purchasing capital-intensive continuous annealing facilities.

The objectives of the contract evolved somewhat as the work progressed, and the results became available; in this regard, the writer wishes to acknowledge at the outset the capable contract monitoring of A.F. Crawley, PMRL-CANMET, whose concern for and interest in the work ultimately grew into full research collaboration.

The major objective of the research has been maintained through the entire period of the contract, that is, to study and understand the causes of limited ductility of batch annealed dual phase steels, in the expectation that such an understanding will lead to the development of improved material properties, especially as they relate to formability. The initial phase focussed on experimental studies of microchemistry of a high manganese steel developed by CANMET, supplemented by numerical modelling of the development of the chemistry and volume fractions of the microconstituents. In a second stage, the flow and fracture properties of the high-alloy batch annealed material were compared with those of two commercially produced continuously annealed dual phase steels, and the differences related to microstructure and microchemistry. The

final phase consists of the development of batch annealing cycles applicable to lean alloy compositions; this development follows naturally from the first two stages, and brings the research to a logical and potentially useful conclusion.

This report will summarize in some detail the results of the initial phase, which was the subject of a full interim report (1). Major emphasis will then be placed on results of the latter phase, which have not yet been formally reported.

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(1) Interim report, DSS Contract 20 SU.23440-0-9191, June 1982.

SECTION I: STUDIES OF MICROCHEMISTRY OF HIGH MANGANESE DUAL PHASE STEELS

In this study, the material developed at CANMET, a steel containing 3%Mn, 0.06% C, and 0.23% Si, was subjected to simulated batch annealing cycles, and specimens for microstructural observation and microanalysis quenched from various points within the cycles. This procedure is summarized in Fig. I-1, where the numbered arrows indicate that specimens were quenched at these times. Observations of the samples then indicated in summary:

1) The as-rolled steel was strongly banded, and this banding persisted through the entire batch annealing cycle.
2) The banding was correlated with manganese microsegregation, wavelength of order 100 micrometers, and amplitude approximately 1% Mn. This information was derived from electron microprobe analysis, Figure I-2.
3) Ultra-microanalysis was performed on this section of the annealed specimens, using the scanning transmission electron microscope, and X-ray spectroscopy. These studies indicated that:
    a) Near-equilibrium partitioning of manganese to the martensite-austenite phase occurred, after relatively short intercritical annealing periods, Figure I-3; and
    b) Carbides were enriched in manganese during subcritical heating to the intercritical annealing temperature, to the point where some carbides survived prolonged intercritical annealing, Figure I-4.

Figures I-3 and I-4 are typical of the numerous microanalytical results obtained from simulated batch processing of the CANMET steel. Both of these observations are at variance with experience pertinent to continuous annealing treatments of low manganese steels (2). Both point to the major influence of subcritical annealing (in this case, during the slow heating to the intercritical tem-

ments of low manganese steels (2). Both point to the major influence of subcritical annealing (in this case, during the slow heating to the intercritical temperature) on the austenite microchemistry, formation kinetics and final properties.

Because of the importance of equilibrium or near-equilibrium partitioning of manganese during subcritical annealing, a series of ternary Fe-C-Mn isotherms were computed as an aid to the design of heat treatments. These are included as Appendix I.

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(2) Interim report, DSS Contract 20 SU.23440-0-9191, June 1982.

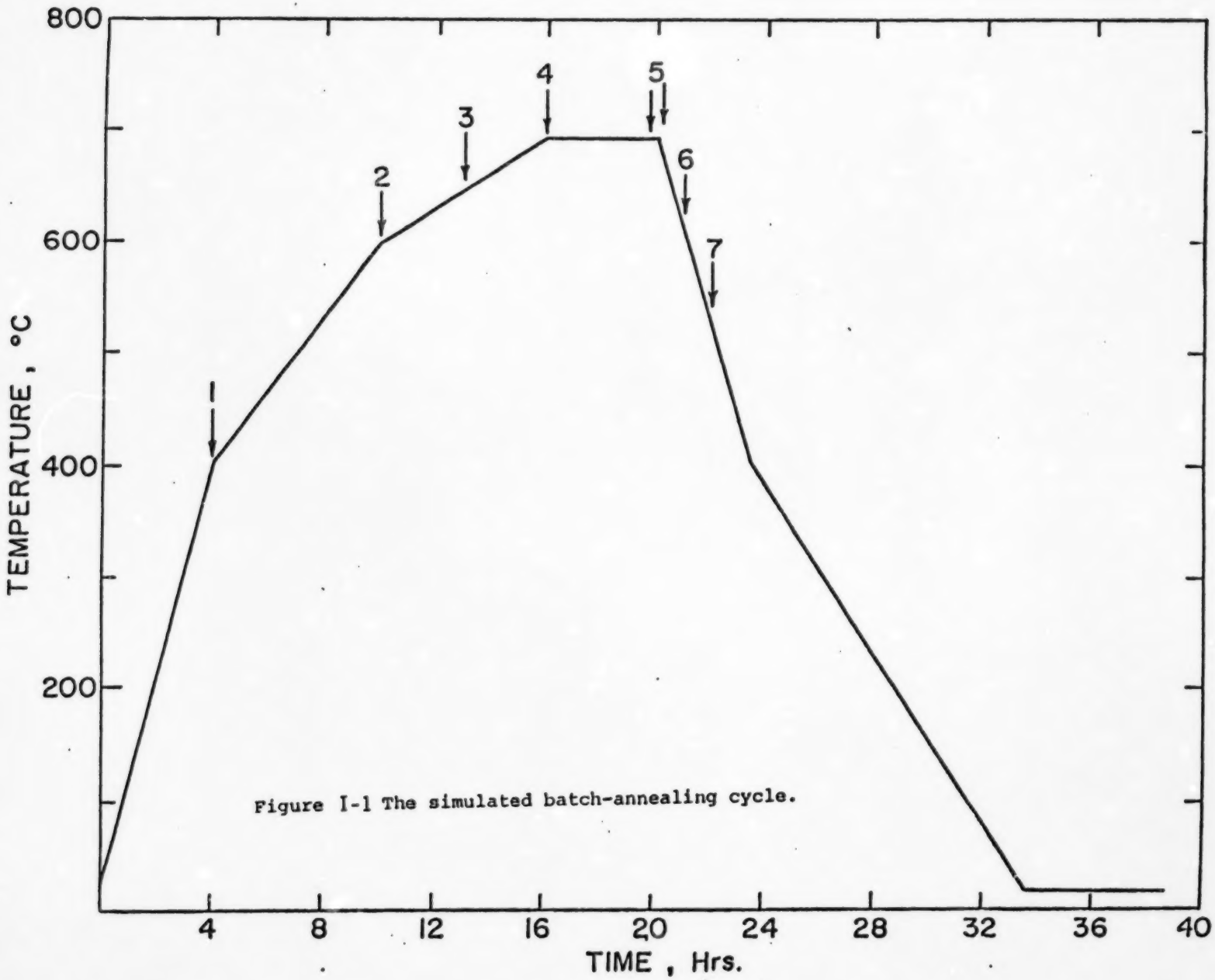

Figure I-1

Figure I-1 The simulated batch-annealing cycle.

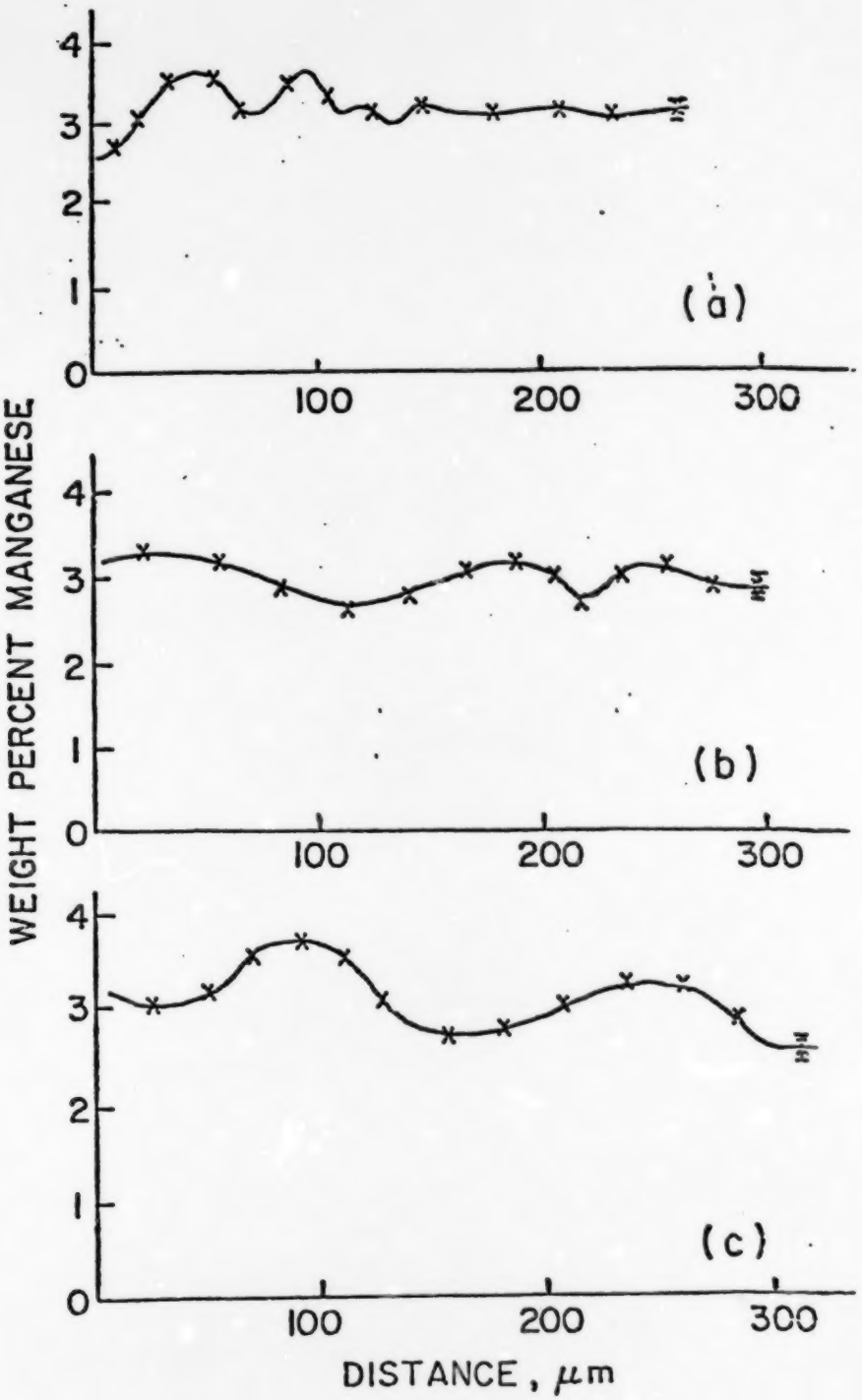

Figure I-2 Microprobe scans of plate material

a) As received
b) Annealed 4 h at 715°C
c) Annealed 4 h at 695°C, cooled 2 h to 525°C.

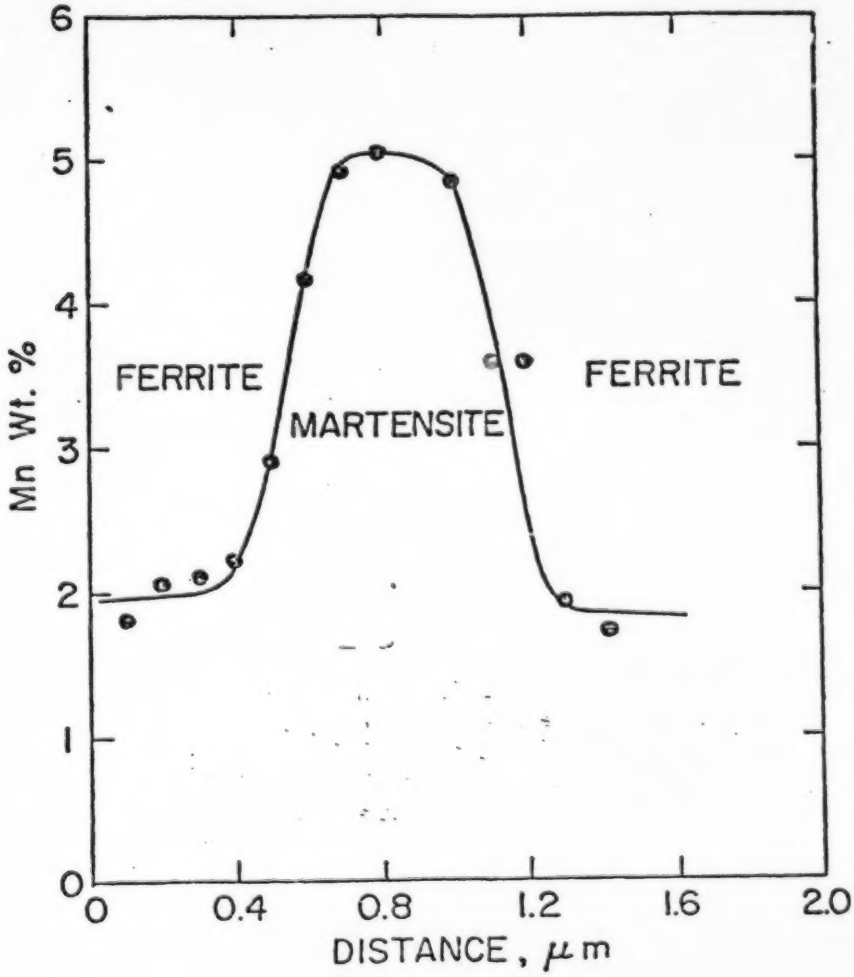

Figure I-3 Manganese profile from a STEM step scan across ferrite/martensite interface.

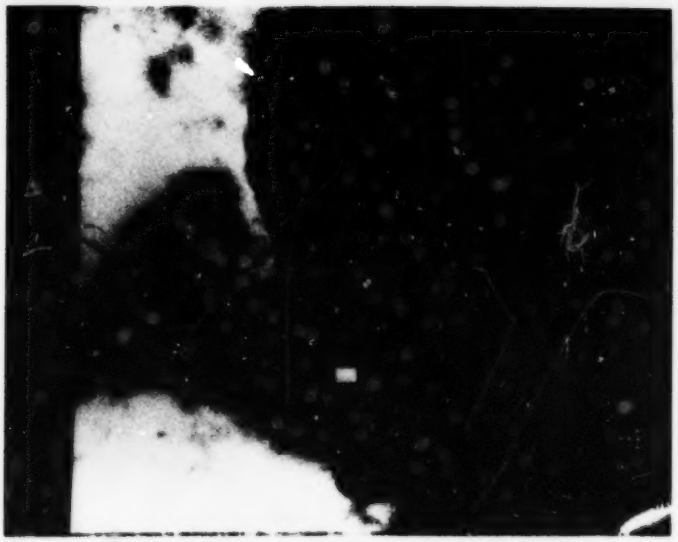

Figure I-4A:
Persistent carbides
in specimen 4.

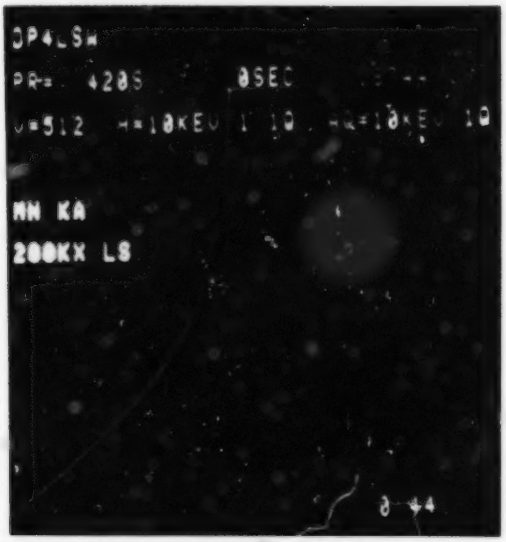

Figure I-4B:
Mn linescan across
carbide of 4(a).

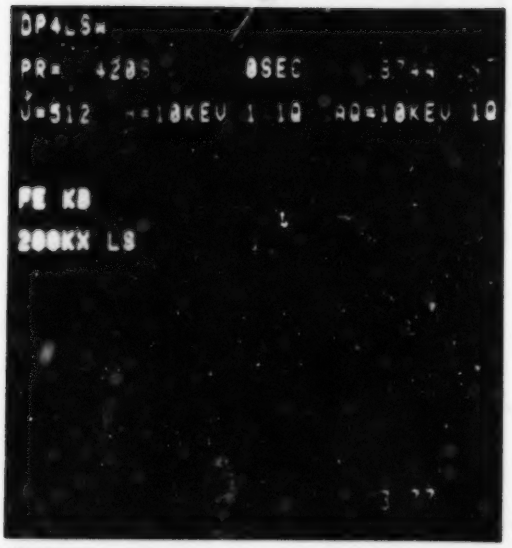

Figure I-4C:
Fe linescan across
carbide of 4(a).

SECTION II: FRACTURE AND DUCTILITY STUDIES OF BATCH AND CONTINUOUSLY PROCESSED DUAL PHASE STEELS

In this section of the work, an intensive metallographic and fractographic investigation was carried out on three materials; two commercially produced continuously annealed dual phase steels, and the CANMET batch annealed steel. The microstructural properties of the three steels were correlated, as far as possible, with the tensile and fracture properties. Prior studies of the high Mn dual phase steels have shown the characteristic dual phase steel properties; rounded yield point, low YS/TS ratio, and high tensile strength can be obtained (1). There have been problems, however, e.g.

(1) Uniform (and total) elongation is lower for a given tensile strength than that of the lower manganese dual phase steels produced by continuous annealing (1,2,3).

(2) Property variation throughout a coil can be expected to be comparatively larger for the high manganese grades.

(3) High manganese dual phase steels, in particular, are prone to severe banding; this is attributed to initial transformation segregation of carbon attending manganese microsegregation. It has recently been shown, however, that banding can be developed or intensified during prolonged intercritical annealing (4).

Previous investigations of continuously annealed dual phase steels have indicated two general trends of interest:

(1) The uniform and total elongation and the work hardening rate all tend to reduce as manganese content is increased (1,5).

(2) Prolonged intercritical annealing is associated in the literature with decreased total elongation (6).

However, the reports of these investigations give no clear indication of the basic reasons for this behaviour. We might expect that the uniform and total elongation will be improved by reducing manganese content. The reasons for the deleterious effects of long-term annealing are more difficult to understand. A fuller understanding of these factors would be of value in designing steels for batch annealing processing.

EXPERIMENTAL:

The chemistries of the steels investigated are shown in Table 1. Steel "C" is a batch annealed dual phase steel, processed as follows:

1) Wrapped in a stainless steel envelope, placed in a cold furnace, and heated to 709°C and held at this temperature for 5 hours.
2) Cooled at 27.8°C/h, then removed at 90°C. The JL and N designations refer to commercially produced steels which were continuously annealed. Steel "N" was cold rolled.

The microstructures of these steels were examined using optical and scanning electron microscopy; observations were made on prepolished surfaces of tensile specimens, and on fracture surfaces, as well as on sections of as-received and deformed material. Tensile tests were carried out with an Instron Universal Tester, operated at a crosshead speed of 1.25 mm/min.

RESULTS:

Tensile Tests:

The engineering mechanical properties are shown in Figure II-1 and Table 2. The high-Mn batch-annealed material "C" has the highest tensile strength and the lowest uniform and total elongation; steel "N", cold rolled and continuously annealed, has the lowest tensile strength and the highest uniform and total

elongation; "JL" which contains vanadium and is continuously annealed, is very similar to "N" in its mechanical properties.

The true stress/true strain curves for these steels are shown in Figure II-2 and their work hardening rates in Table 3. The initial work hardening rate (n1) of steel "C" is the highest, but in high strain regime, it falls significantly; "N" shows the lowest initial rate while "JL"'s is close to that of "N".

Microstructural Observations:

The as-received microstructures of these steels are shown in Figure II-3, and some microstructural parameters of these steels are listed in Table 4. There are some noteworthy differences in the microstructure of these steels.

The morphology of the matrix (ferrite) is quite different in the continuously processed and batch annealed materials. The matrix of steels "N" and "JL" consists of polygonal ferrite, but the high Mn "C" possesses a structure as acicular as that of upper bainite (the second phase, austenite at annealing temperature, also occurs in acicular or plate form). It is likely that the acicular microstructure is formed on initial cooling from the finishing temperature, and retained through the intercritical anneal. This view is supported by observations of steel "C" as hot-rolled (Figure II-4). It can be seen from Figures II-3C and II-4 that the morphology of the ferrite and second phase is similar in the two conditions. A similar observation was made by Shehata and Crawley, in their studies of similar materials (7).

The second-phase morphology for steel "N" is also polygonal. There are many dark regions at the boundary of the second phase; these regions may be martensite of higher carbon content. In the centre of the second phase, the martensite probably has a lower carbon content, due to carbon gradients formed in the austenite during the growth of epitaxial ferrite on cooling. In some areas,

bainite (bainitic ferrite and high carbon martensite) can be seen (Figure II-3a).

The deformation behaviour of the three steels was characterized by microstructural observation, as well as by the tensile results of Table 2; in this part of the study, surface slip line behaviour, and the microstructures of polished sections of the necked regions were characterized, and compared with the appearance of the fracture surfaces themselves.

The studies of surface slip lines are summarized in Figure II-5. The most notable distinction between the continuously processed and batch annealed materials lies in the persistence of coarse slip up to the second phase particle boundaries in the former case, and the restriction of coarse slip to the centres of the ferrite regions for batch processed steel. This suggests a higher degree of plastic constraint, exercised by the more nearly continuous second phase layers in steel "C"; this is discussed later.

Metallographic observations of the high strain and fracture regions of the tensile specimens yielded further distinctions among the materials. In all cases, the microstructures contain the suggestion of (limited) plastic deformation in the non-ferritic phase (Figure II-6). The initiation of fracture appeared to be quite different for the two types of material, however, in that the voids nucleated at the ferrite-martensite interface in continuously annealed steel, whereas the elongated martensite regions in steel "C" were often found fractured, (Figure II-7). In spite of this major difference in void nucleation mechanism, the estimated strain required for void nucleation was similar in all cases (Figure II-8). The significant difference rests in the strain required for the growth of the voids.

The final fractures were all fully ductile, as expected. The fractographs of Figure II-9 and the fracture cross-sections of Figure II-10 illustrate the lesser spacing between voids, and the reduced ductility characteristic of the batch annealed steel "C", as compared with the two continuously annealed steels. The

quasi-shear fracture also appeared to be characteristic of the batch annealed steel, and contrasts with the more typical cup-cone fracture found for the other materials.

DISCUSSION:

Our main objective is to discern the main reason or reasons for the reduced ductility and the fracture characteristics of steel "C". Since both the chemistry and the microstructure of this material are different from that of the comparison materials, it is not immediately apparent which of the distinctions is the more important. We will begin by considering the microstructural differences.

The microstructural differences are associated, in the main, with the scale of the structure, and with the degree of constraint exercised by the heavily banded acicular nonferritic constituent*. This constraint, applied to the ductile phase, might be expected to limit its capacity for plastic deformation. In the limit of a hypothetical structure composed of alternating contiguous slabs of martensite and ferrite, we would clearly have a composite whose ductility was limited by that of the martensite. It is particularly apparent that the void nucleation process, while occurring at similar levels of strain in the two cases, involves quite different fundamental processes. In the case of the equiaxial structures, void formation occurs by decohesion at the ferrite-martensite interface; in the banded material, the martensite plates are fractured, and plastic strain in the ferrite will therefore be concentrated in the regions adjacent to the fracture.

It is probable that the quasi-shear failure mode also results from the plate-like banding that is so pronounced in steel "C". The model of Figure II-11 repre-

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* We will call the second phase martensite in subsequent discussion, although this is clearly a simplification.

sents martensite slabs in a ferrite matrix. When this material is loaded in tension, the elastically deforming martensite regions carry a much higher stress than the (plastic) ferrite layers. A crack nucleated in the martensite at A will therefore propagate throughout the martensite on the horizontal plane containing A. The long slit-shaped crack will transfer its load to the surrounding ferrite, and depending on the slab spacing, to the neighbouring martensite layers, causing a second crack at B. If a second crack is not formed, further strain of the specimen as a whole will tend to open the crack at A, raising the surrounding strain levels and increasing the probability of a second crack nearby. Since the martensite and ferrite have similar elastic properties, the state of stress created by long thin cracks at A and B is comparable to that treated by Isida (8). If A and B are not coplanar, a shear stress component exists on the line joining them.

If the angle between the plane joining the two cracks and the horizontal is greater than 30°, and if the general stress level is at the ferrite flow stress, it is credible that a shear band will develop between A and B. This will displace the crack faces and propagate the shear band into the adjacent ferrite layers. When the shear band reaches the next martensite slab at C, the local shear strain concentrated at the tips of the band will cause a shear failure in these martensite slabs, allowing further stretching and shearing between A and B. Continuation of this process would lead to a slanting type of fracture surface of the type observed in type "C" batch annealed steel.

The most important factor differentiating this type of failure from the relatively isotropic void linkage found for the other two materials is that the initial fracture in the slab is a long thin slit. The difference in the stress intensity factor along the long edge of the crack compared with that at the end of the crack creates a plane strain condition along the slit when only one crack exists. As the second and subsequent cracks stretch out and create free surface, this will

change to a plane stress condition in the ferrite. In either extreme, the preferred movement of material will lead to the slant fracture.

Thus, we can understand, at least semi-quantitatively, the differing void nucleation and fracture characteristics of the two types of steel, and relate these differences to gross microstructural characteristics.

Some further insight into the deformation characteristics of these steels may be obtained from an analysis of the stability of plastic flow, as shown in Figure II-12.

In Figure II-12, the steels "C" and "N" are compared in terms of their work hardening characteristics; the limit of uniform elongation is determined by the intersection of the flow curve and the work hardening curve in each case. The major difference clearly lies in the higher rates of initial work hardening displayed by the batch annealed steel "C", before the initiation of fracture within the martensitic constituent (at a strain of 0.08). This high initial rate increases the strength of steel "C" so that the point of plastic instability is reached at a lesser strain.

It is not immediately clear whether this high initial rate of hardening is related solely to the degree of constraint exercised by the martensite on the ductile ferrite, and thus to the platelike morphology of the high manganese bands, or whether the microchemistry of the constituent phases (primarily the ferrite) also plays a role in determining the hardening rate.

The phase compositions of Table 4 are crude estimates, based on limited microanalytical results (1), but are thought to be a reasonable basis for discussion. We expect that manganese will be strongly partitioned to the martensite in batch annealed steel, and only weakly partitioned on continuously produced material (10,11). Thus, the manganese levels in ferrite are likely to be similar for steels "C" and "N", while the silicon contents differ substantially. However, "C" and

"JL" have similar silicon contents in ferrite. There appears to be no consistent microchemical trend to relate to the higher initial rate of work hardening, and we are therefore inclined to assign the greater weight to those microstructural characteristics which evidently exert a strong influence on the later stages of deformation.

Note, however, that it is the gross microstructural features associated with severe banding with which we associate the major part of the loss of ductility, and that severe banding is frequently associated with high levels of manganese.

CONCLUSIONS:

These studies suggest that the superior ductility of the continuously processed steels is linked mainly to microstructure. The continuously processed materials possess the equiaxed ferrite-martensite structure typical of dual-phase steels. In contrast, the high manganese batch processed steel is heavily banded. As noted in section I, the banding is closely related to the high levels of manganese, which were added to the steel to give the austenite the required hardenability. The banding is sufficiently severe that the work hardening and fracture nucleation behaviour of the manganese steel is dominated by it. Fracture is nucleated as the cracking of a martensite slab, rather than as the interfacial voiding of an equiaxed martensite island. Thus, it concluded that the design of a dual phase steel for batch processing should not rely on high GLOBAL values of the alloying elements for the necessary austenite hardenability; to duplicate the properties of continuously processed steels, one should start with a lean alloy composition, and selectively enrich the austenite in the desired alloying elements by design of heat treatment cycle. The test of this concept is dealt with in the next section.

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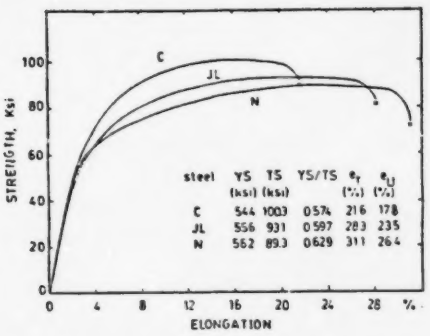

Figure II-1 The engineering mechanical properties of experimental materials.

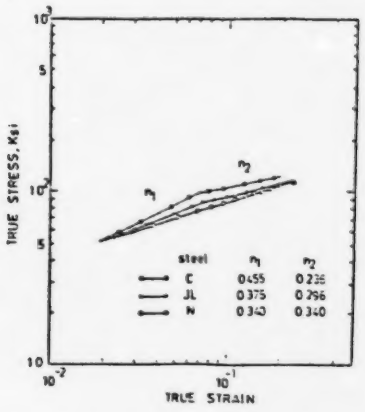

Figure II-2 The true stress/true strain curves of experimental materials.

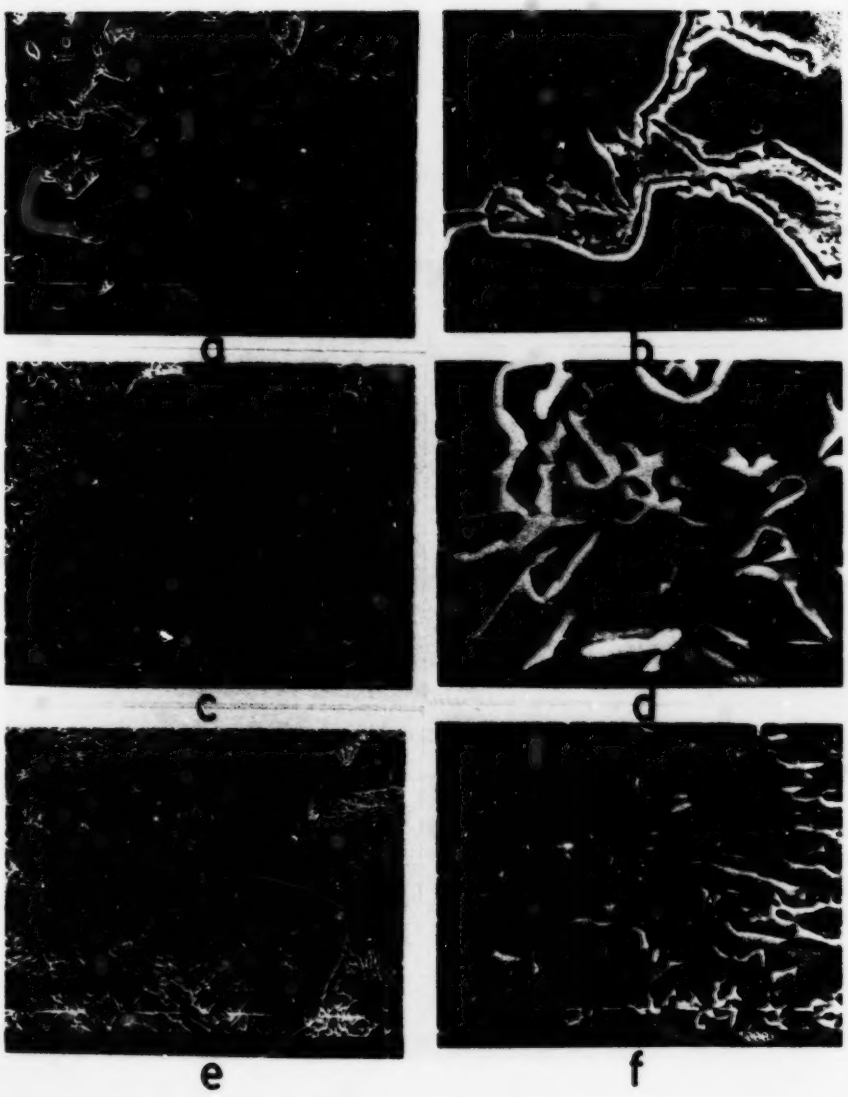

Figure II-3 The as-received microstructure of experimental materials.

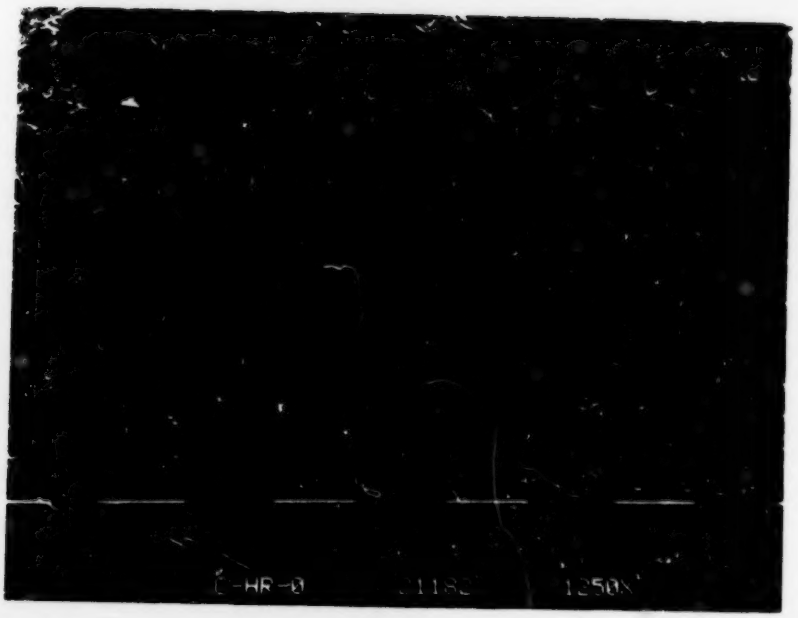

Figure II-4 The as hot-rolled microstructure of steel "C".

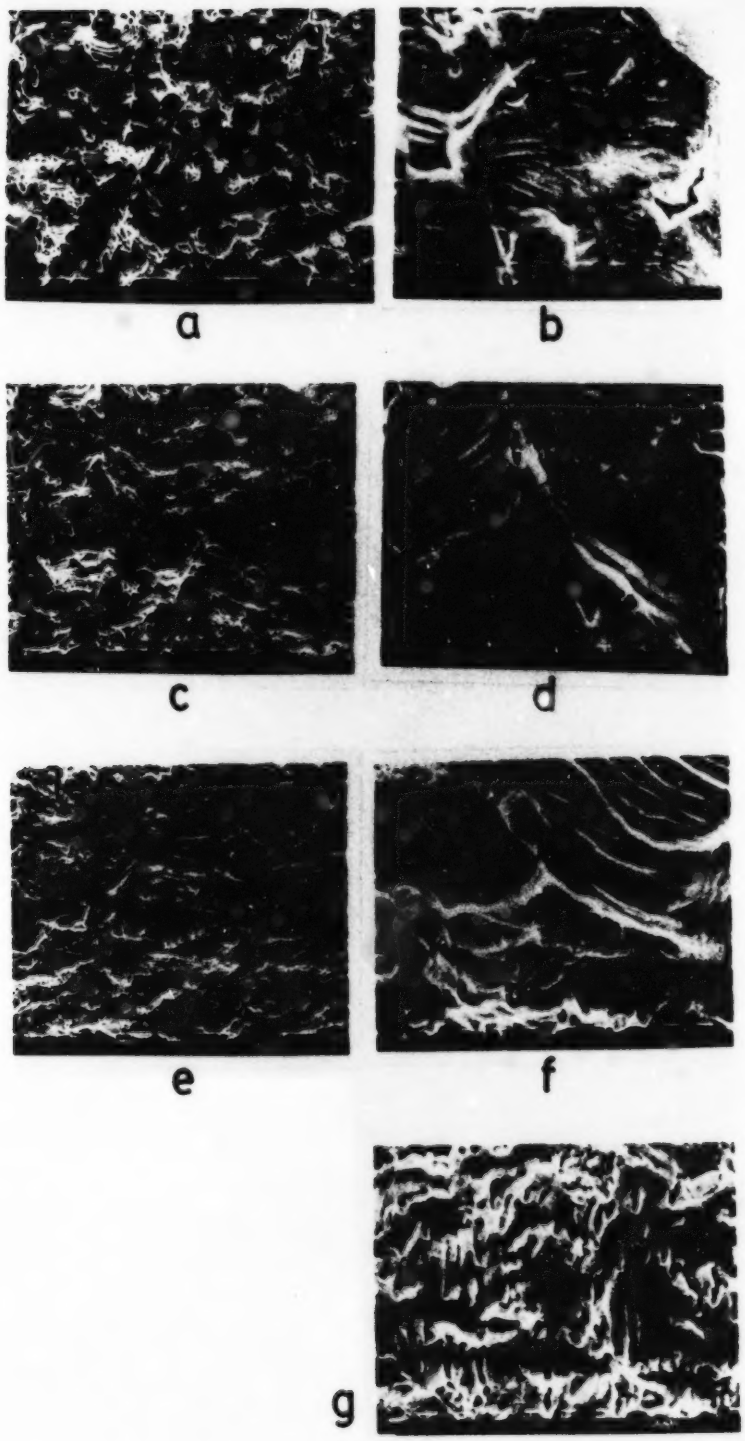

Figure II-5 The surface slip lines of experimental materials.

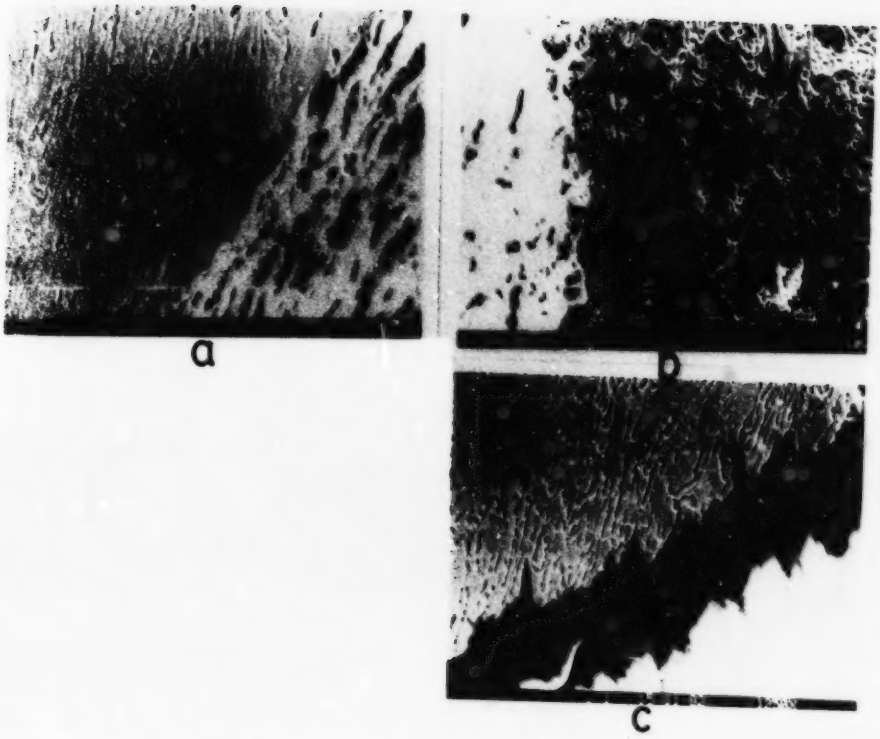

Figure II-6 The microstructure contains the suggestion of (limited) plastic deformation in the non-ferritic phase.

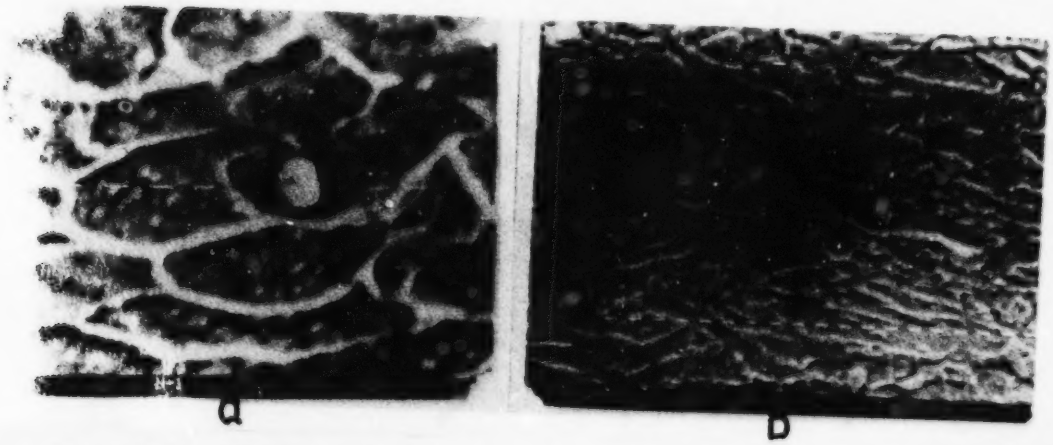

Figure II-7 The initiation of fracture appeared to be quite different for the two types of experimental materials.

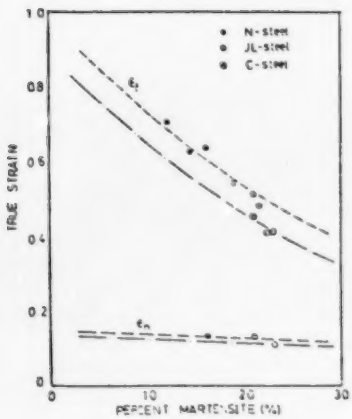

Figure II-8 The estimated strain required for void nucleation and fracture of experimental materials.

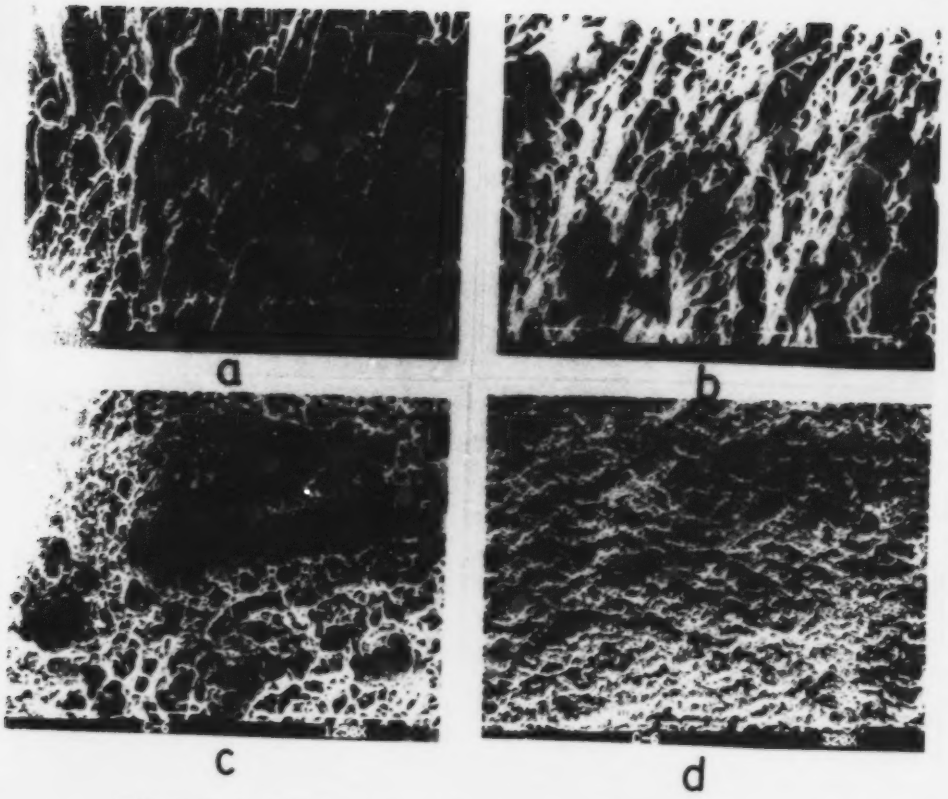

Figure II-9 The fractographs of experimental materials.

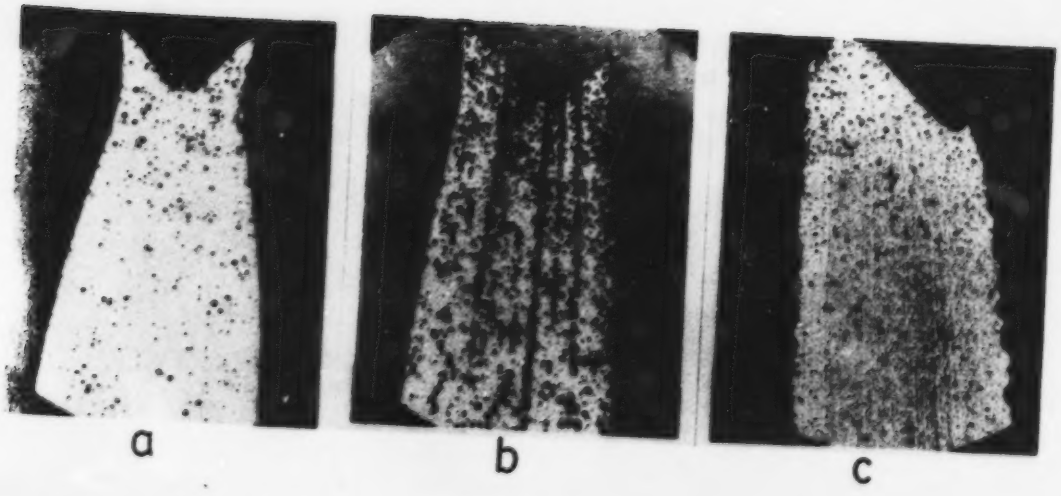

Figure II-10 The fracture cross-sections of experimental materials.

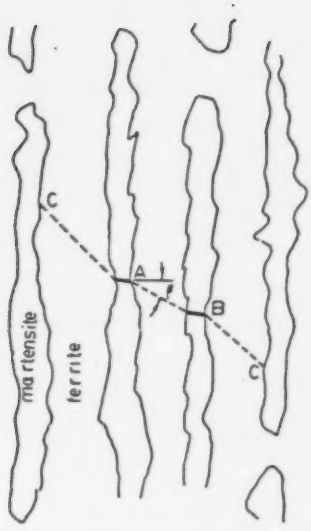

Figure II-11 The quasi-shear failure mode results from the plate-like banding, which is so pronounced in steel "C".

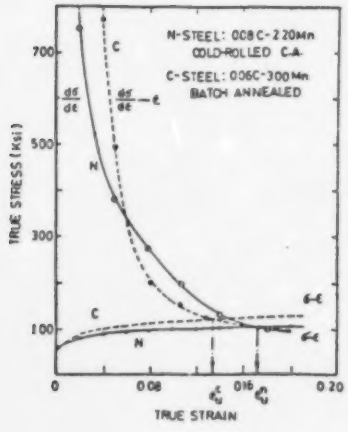

Figure II-12 The deformation behaviour of experimental materials.

TABLE 1 Chemistries of Experimental Steels

Steel	C	Mn	Si	V	Al
C	0.06	2.88	0.31	---	0.06
JL	0.12	1.66	0.57	0.06	0.06
N	0.08	2.20	0.03	---	0.008
C-484	0.11	1.68	0.54	---	0.04

TABLE 2 Engineering Mechanical Properties of Experimental Steels

Steel	YS (MPa)		TS (MPa)		YS/TS		e (%)		e (%)	
		376		692		0.542		21.6		17.8
C	(345)	316	(705)	708	(0.489)	0.447	(25.3)	28.5	(17.2)	18.6
		341		714		0.479		26.5		16.2
		383		643		0.597		28.4		23.5
JL	(375)	376	(654)	665	(0.572)	0.567	(30.2)	30.5	(22.2)	21.6
		362		657		0.551		31.8		21.6
		388		616		0.629		31.1		26.4
N	(394)	402	(629)	621	(0.623)	0.648	(31.5)	28.0	(22.9)	19.5
		385		659		0.596		35.5		22.8

TABLE 3 Work Hardening Behaviour of the Materials Studied

Steel	Work Hardening Rate	
	n1 initial (less than 8% strain)	n2 high strain
C	0.455	0.235
JL	0.375	0.295
N	0.340	0.340

TABLE 4 Microstructural Parameters and Estimated Phase Compositions of the Steels Studied

Steel	ferritic constituent morphology	mean diam. (um)	Estimated Composition, (wt %) Mn	Si	C
C	acicular	---	2.5	0.6	---
JL	polygonal + acicular	6.3*	1.6	0.5	---
N	polygonal	6.8	2.2	0.1	---
	Second phase				
C	acicular (in plates)	---	6.0	0.2	0.35
JL	polygonal	0.8	1.6	0.5	0.5
N	polygonal	3.7	2.2	0.1	0.5
	Volume Fraction of Second Phase				
C	0.21 + 0.02				
JL	0.19 + 0.02				
N	0.12 + 0.02				

*Note that the distance given here is a measure of the largest ferrite grain diameters, and not a full description of the scale of the structure.

SECTION III: DESIGN OF A HEAT TREATMENT CYCLE FOR BATCH ANNEALING PROCESSING OF LEAN ALLOY DUAL PHASE STEELS

INTRODUCTION:

Following the findings detailed in sections I and II of this report, two possible processing routes were considered for the enhancement of alloying element concentration in the austenite. The principle underlying each of these has been demonstrated by modelling studies, and experimentally confirmed by microanalytical studies (1,2,3,4). In this phase of the work, both mechanisms were considered possible candidates for the basis of a process design. The first is the progressive enrichment of the outer rim of the austenite particle, as growth takes place under conditions of local interfacial equilibrium during prolonged intercritical annealing. Thus, for example, a 1% Mn steel might develop an enriched outer rim on the austenite particles which reaches 3% Mn at its highest concentration (2). Given the strong influence of the manganese concentration on the stability of austenite against diffusional decomposition and the relative importance of the interfacial concentration, it seemed reasonable to attempt the design of a processing route on our modelling studies of this effect.

A second effect which can result in an enriched austenite phase was demonstrated in Phase I, and documented more completely in a concurrent Master's thesis (4). This is the strong tendency for manganese (and other carbide stabilizing elements) to concentrate in the carbide phase on prolonged subcritical annealing. Thus, the slow heating rates inherent in batch annealing processing might be exaggerated and put to use in the enrichment of the precursor phase for austenite in alloying elements. This concept was also put to experimental test, and proved the more successful. Only the results obtained using this latter concept are given here.

EXPERIMENTAL:

The candidate material chosen was one of the commercial continuously annealed steels investigated in Phase II. It is designated JL, and contains 1.66% Mn, 0.12% C, 0.06% V and 0.5% Si. The material was supplied as cut tensile blanks, already possessing a dual phase microstructure. The tensile specimens were sealed in vycor capsules and annealed for one hour at 870°C in order to eliminate the dual phase structure. They were then subjected to processing cycles based on either of the two principles discussed above. In all cases, heating and cooling rates typical of batch annealing furnaces were employed. In all cases, the thermal cycle is shown graphically with the resulting tensile response.

RESULTS:

The results are summarized in Figures III-1A through III-4A, which show representative microstructures resulting from cycles which contain a period of subcritical annealing prior to the intercritical treatment. Figures III-1B through III-4B are composites which show the heat-treatment cycle and the resulting tensile response. In all cases, a small amount of yield point elongation resulted from the thermal cycles. Therefore, certain of the cycles include a final cold rolling stage to demonstrate that this effect may be eliminated, if this is thought necessary.

DISCUSSION:

Of the two suggested processing methods, it was clear upon examination of the microstructures and the tensile results that the second, involving a prolonged subcritical annealing stage, was the more successful. By this means, it is possible to generate very attractive combinations of tensile strength (90 to 100 KSI), and ductility (36% to 30%), with minimal yield elongation. In this respect, Figures

III-4A and III-4B are especially noteworthy. From the microstructures, it is clear that the majority of second phase is indeed martensite, although some minority second phase is present in all specimens as attested by the yield elongations in all as-annealed tensile curves. The reason for the stability of the austenite in the subcritically processed specimens is clearly the concentration of manganese and vanadium in the carbide precursor, as suggested in the report on Phase I of this work, and demonstrated in more detail in reference (4). Thus, the austenite inherits from the carbides a high alloy content, and this contributes to its "micro-hardenability". It is important that the intercritical annealing period, then be kept as brief as possible, consistent with dissolution of most of the carbides, in order that the growth of austenite be limited, and that its carbon content as a consequence remain high. Thus, the intercritical periods are chosen as 1.5 to 3 hours.

As demonstrated by the cycles, including a final rolling pass, it is possible to remove the small yield elongation (Figures III-5 to III-7). It may not be necessary, or even desirable to do this, however, as recent experience has shown that a small amount of yield elongation is beneficial in suppressing the tendency to wrinkle in forming processes (5). If, however, properties corresponding more closely with those associated with continuously annealed steels are required, they can be generated with the aid of a "skin pass", involving .75% to 1.5% deformation.

It should perhaps be emphasized that the present results were obtained with a lean alloy designed to be processed by continuous annealing. Therefore, the results, as they are specific to this alloy, cannot be obtained with other, similar materials without a certain amount of preliminary experimentation. It is expected, however, that the method is sufficiently general that slight variations on the thermomechanical cycles used here will be sufficient for the attainment of similar properties in other compositions, if this is required.

CONCLUSIONS:

1. It has been found that the general characteristics of dual phase steels, that is, good combinations of tensile strength and ductility, can be generated in lean alloy steels, processed by batch annealing cycles.
2. These tensile properties are associated with good sheet forming characteristics. The current materials have a small yield point elongation, which may be beneficial. If not, it can be eliminated by a light rolling pass.
3. The key to the stabilization of austenite against diffusional decomposition, and the consequent formation of martensite as the major second phase, is the concentration of alloying elements in the carbide, using a prolonged subcritical annealing period, followed by a short intercritical anneal. This principle should be applicable to a wide range of lean alloy steels, all of which are now possible candidates for batch annealing processing.

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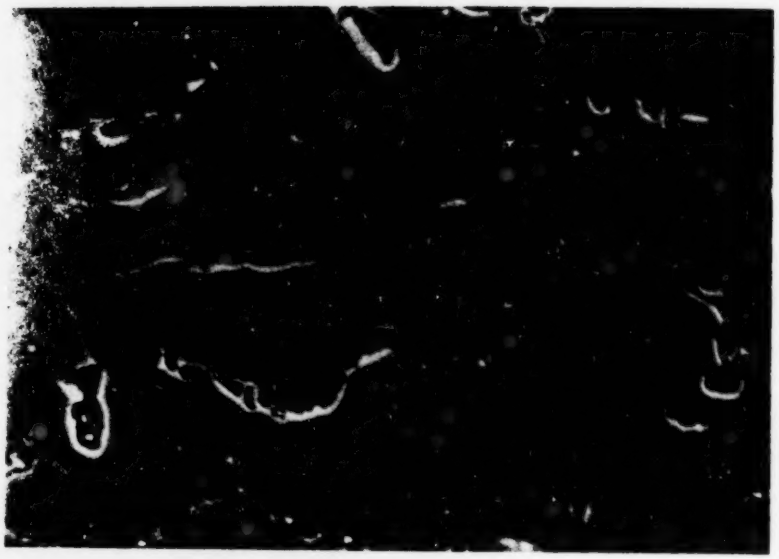

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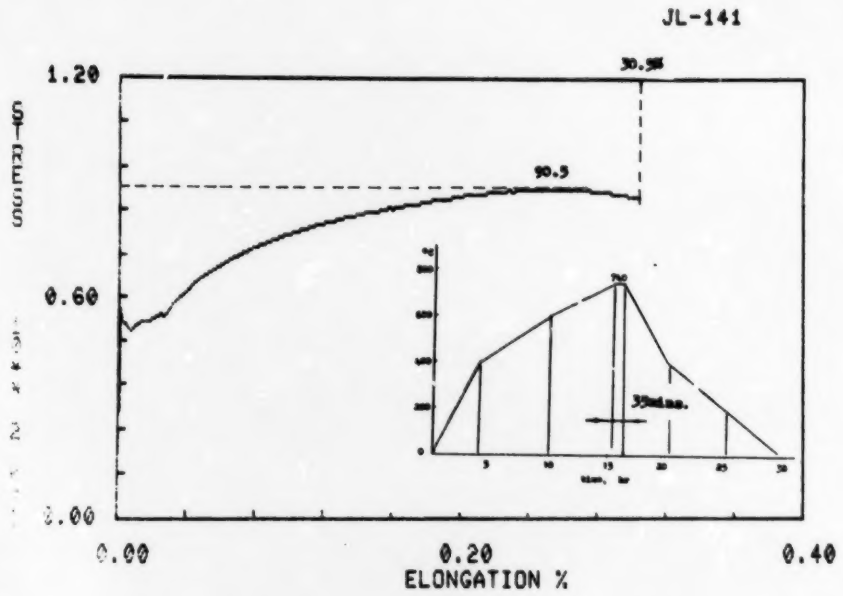
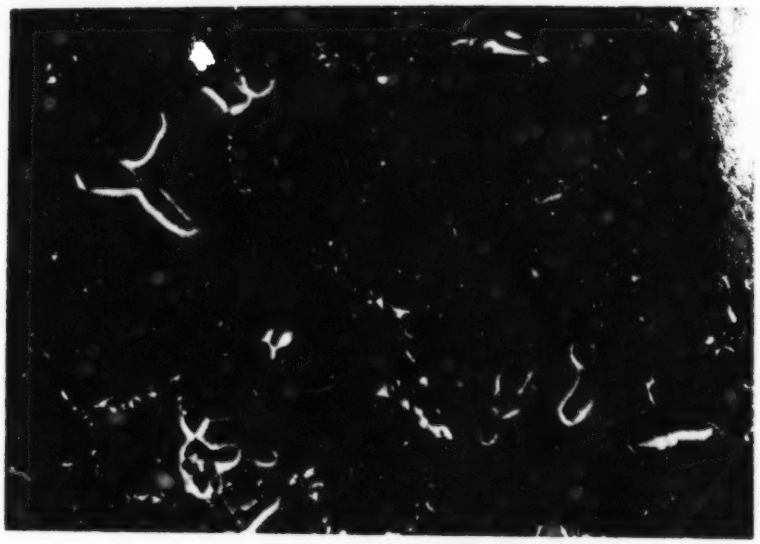


Figure III-1B



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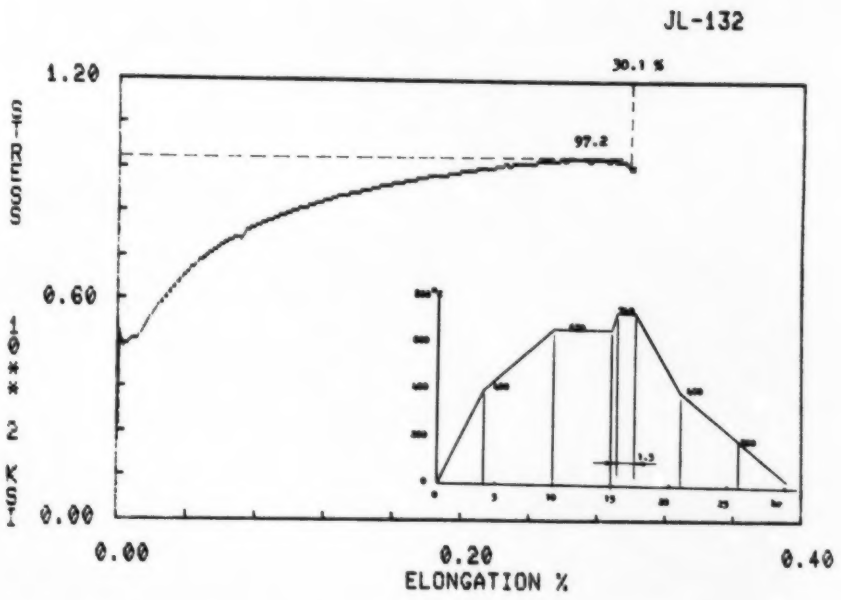
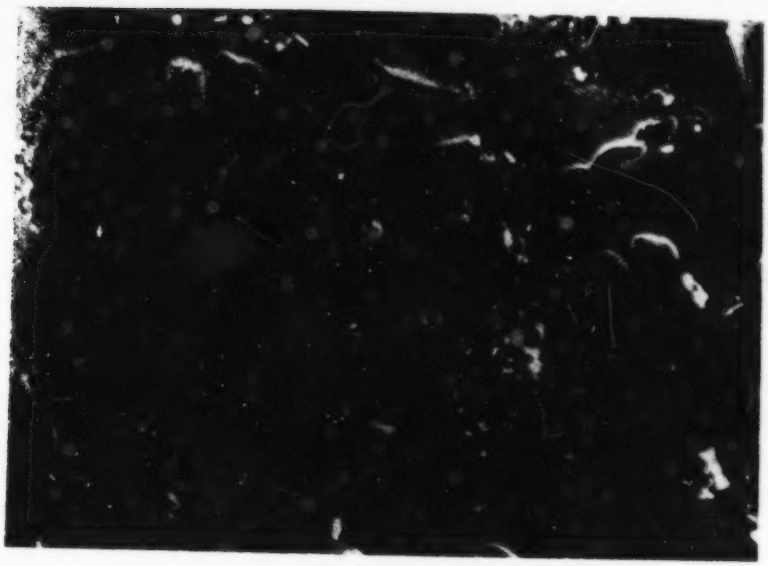


Figure III-2B



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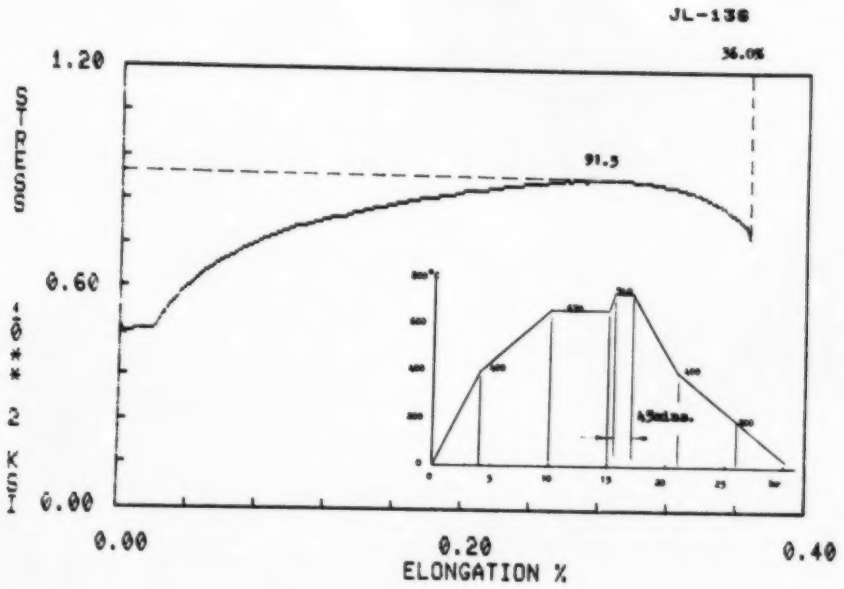
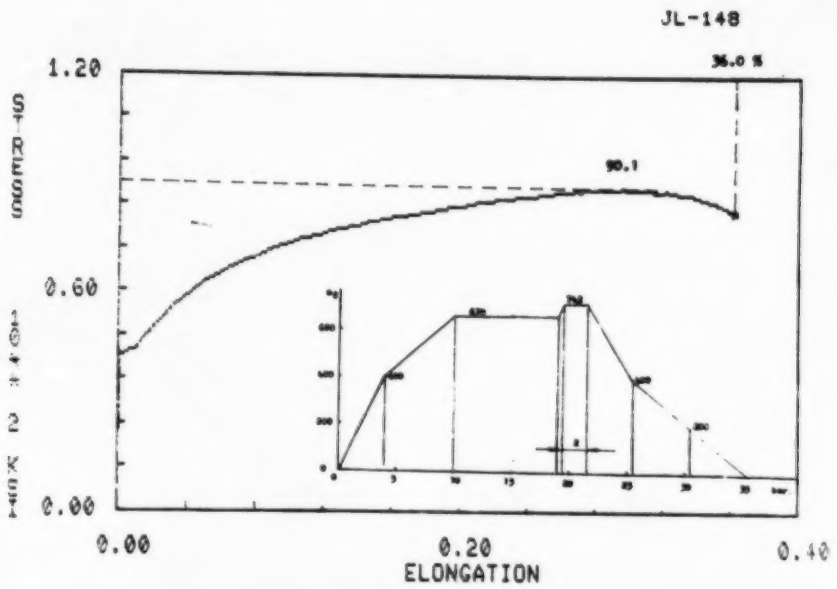


Figure III-3B

Figure III-4A 2500X



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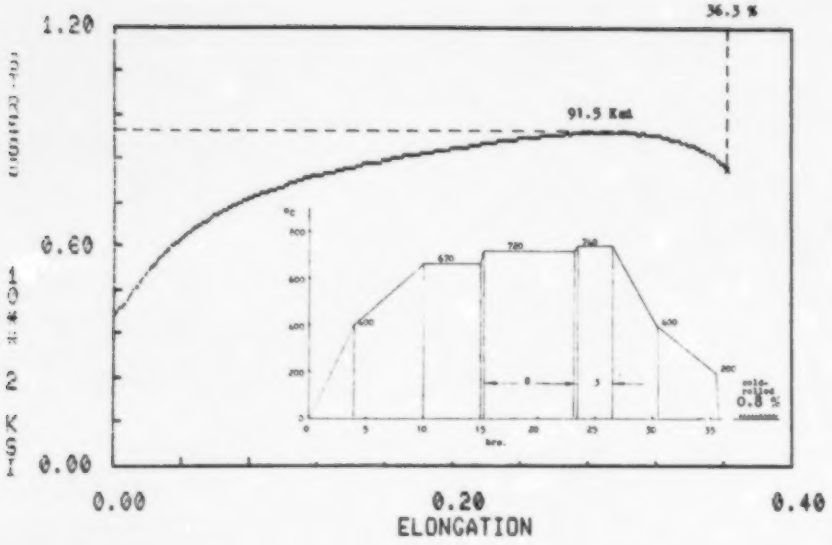
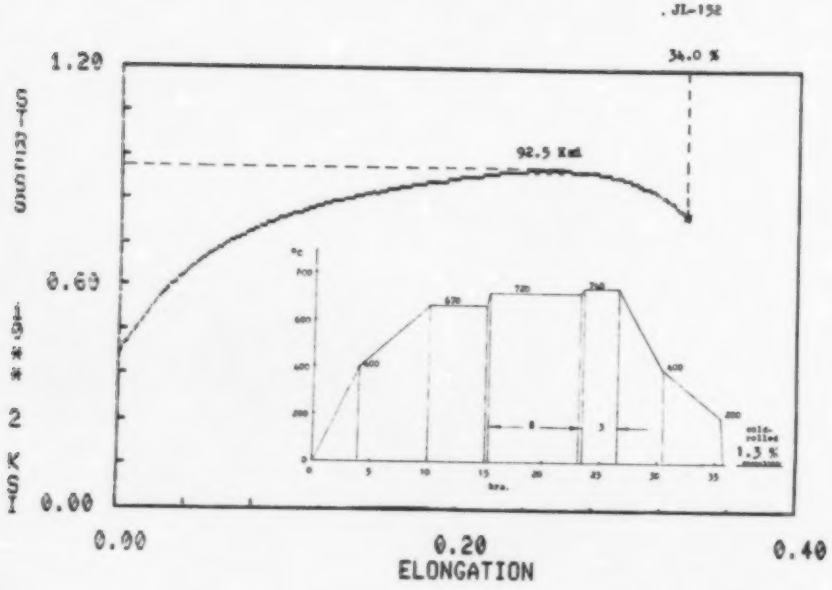


Figure III-5



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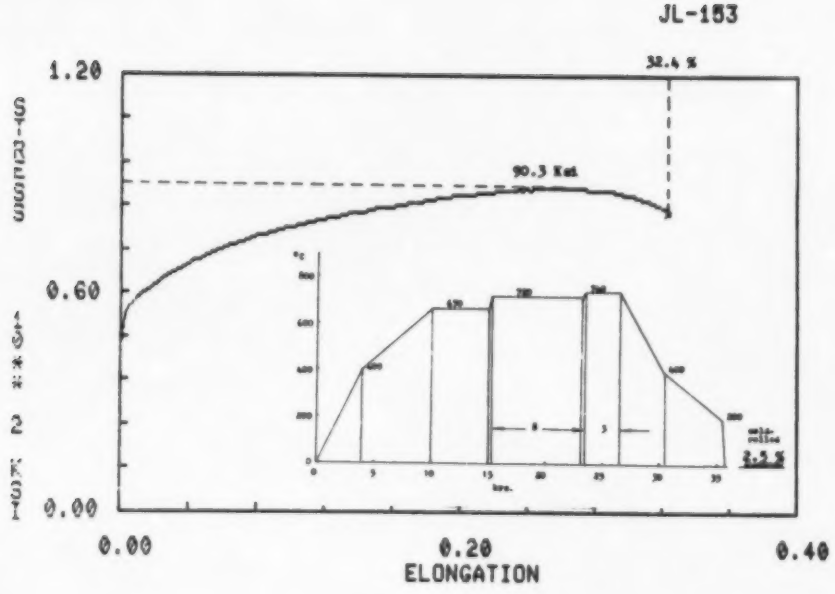
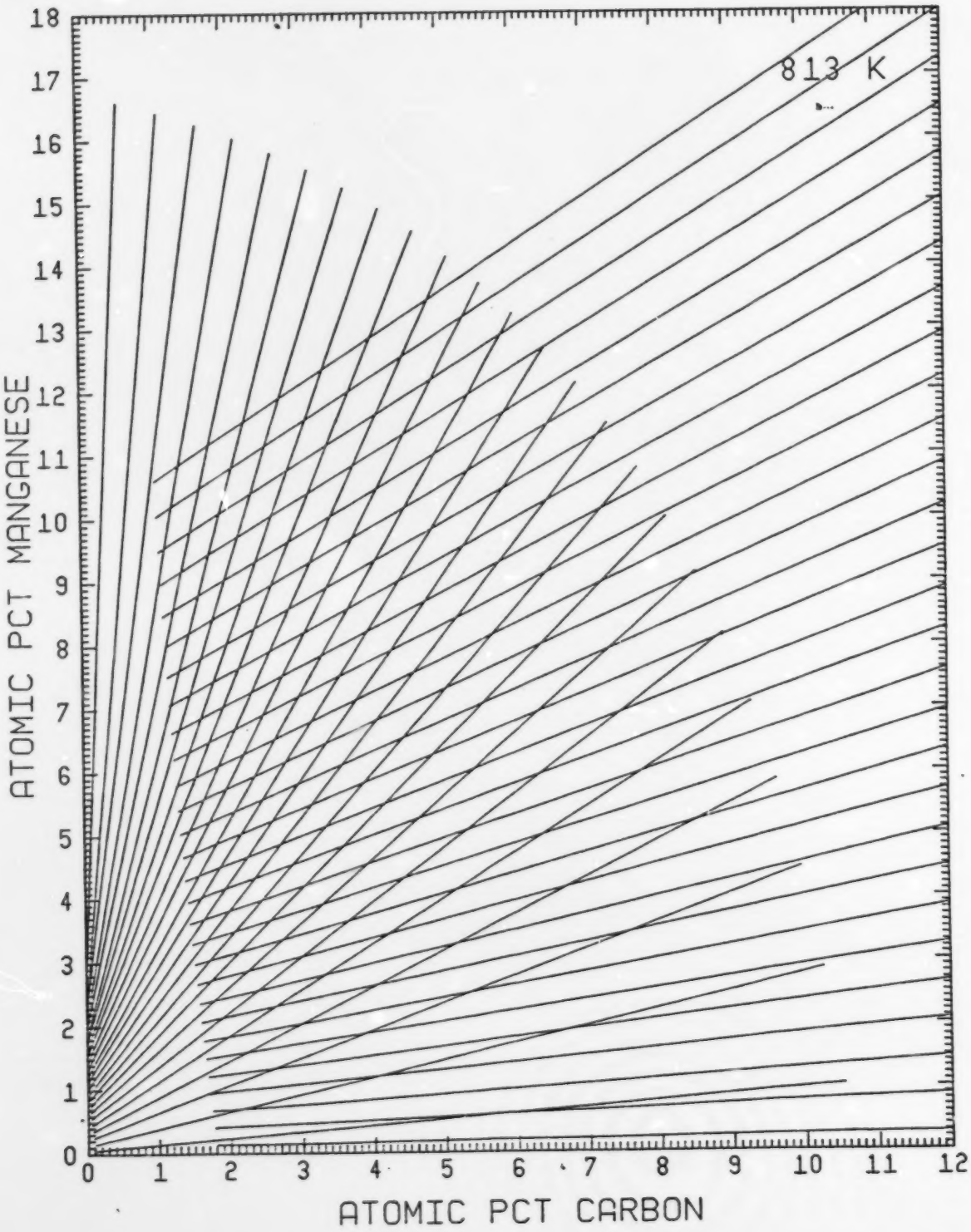


Figure III-7

APPENDIX I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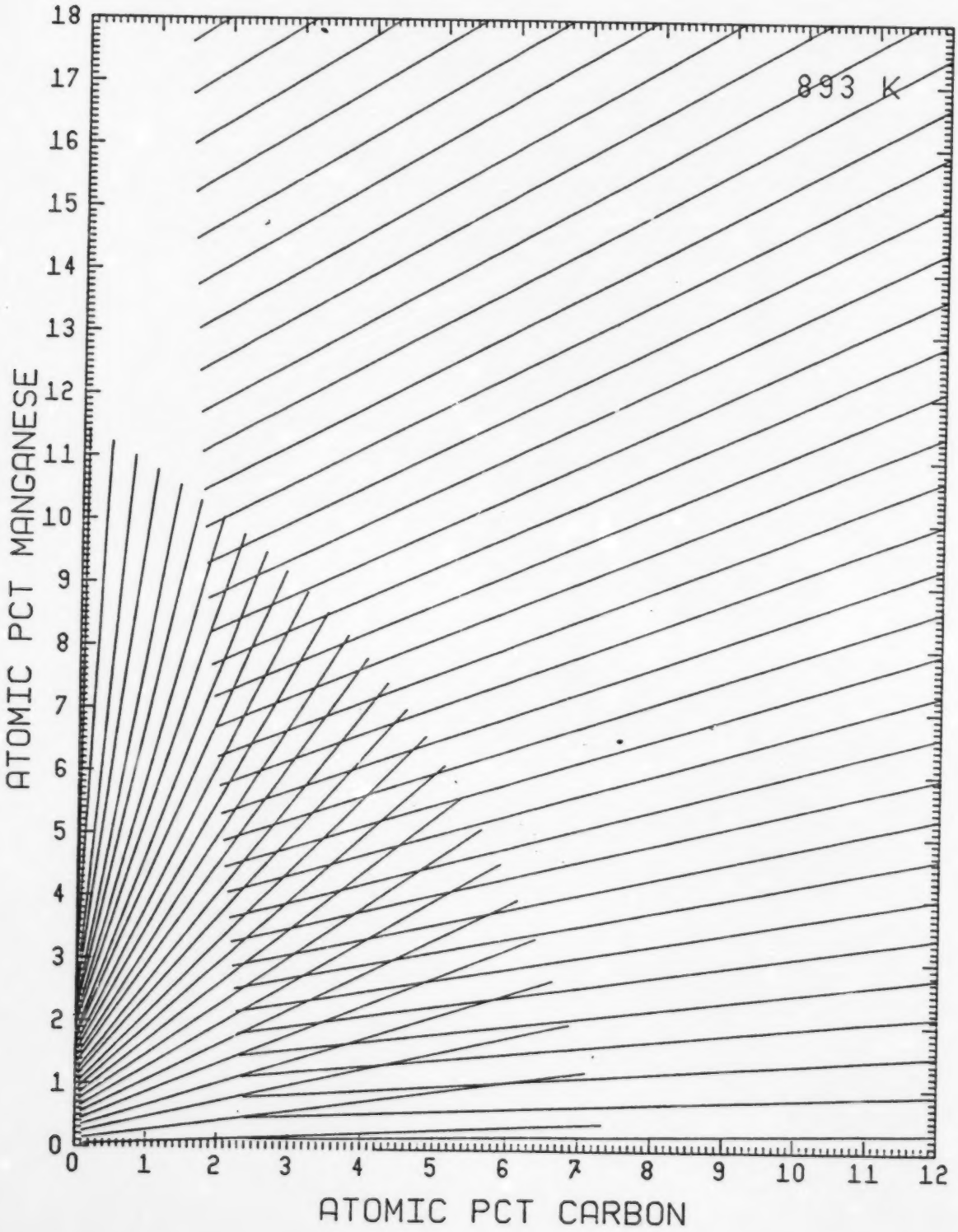
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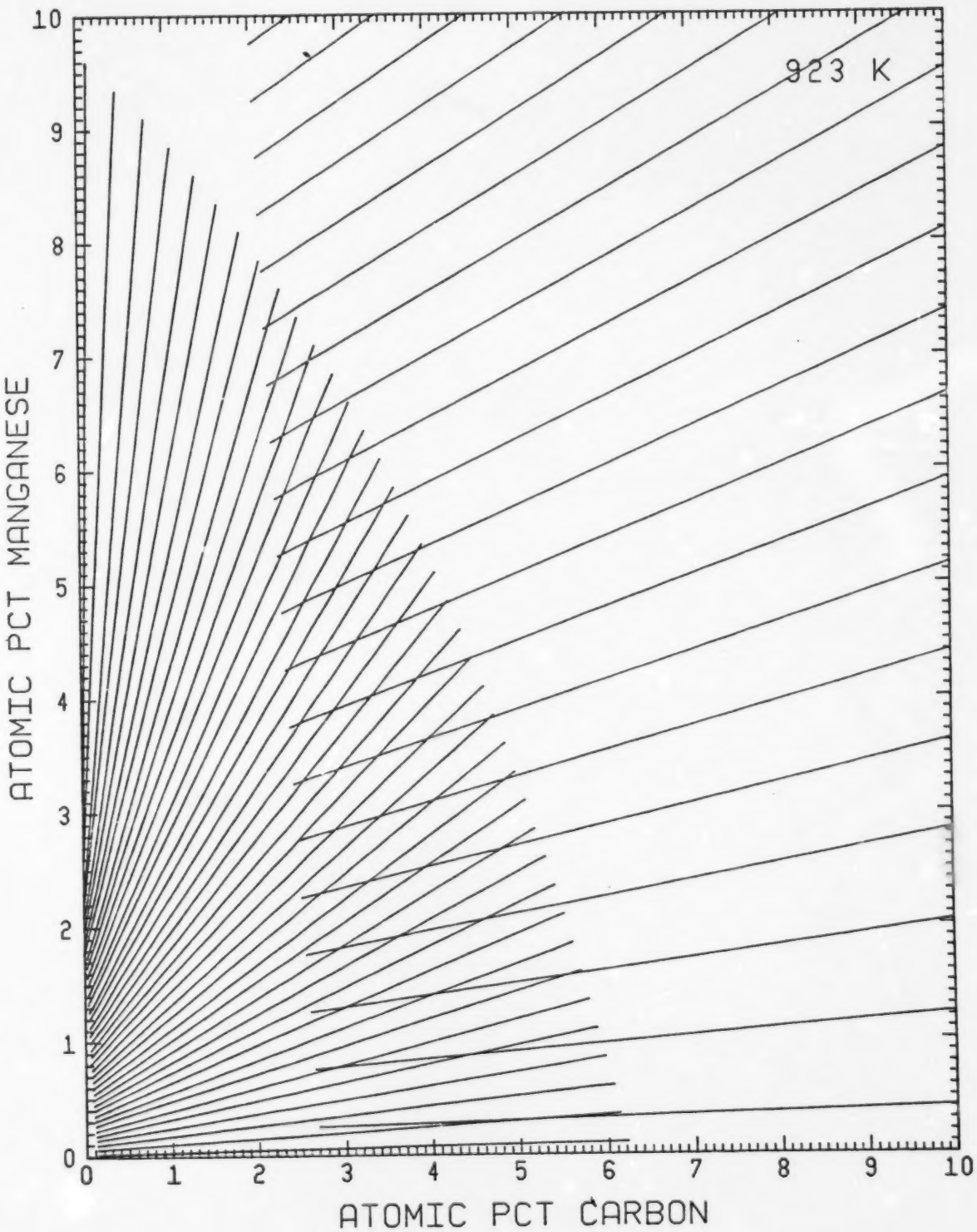
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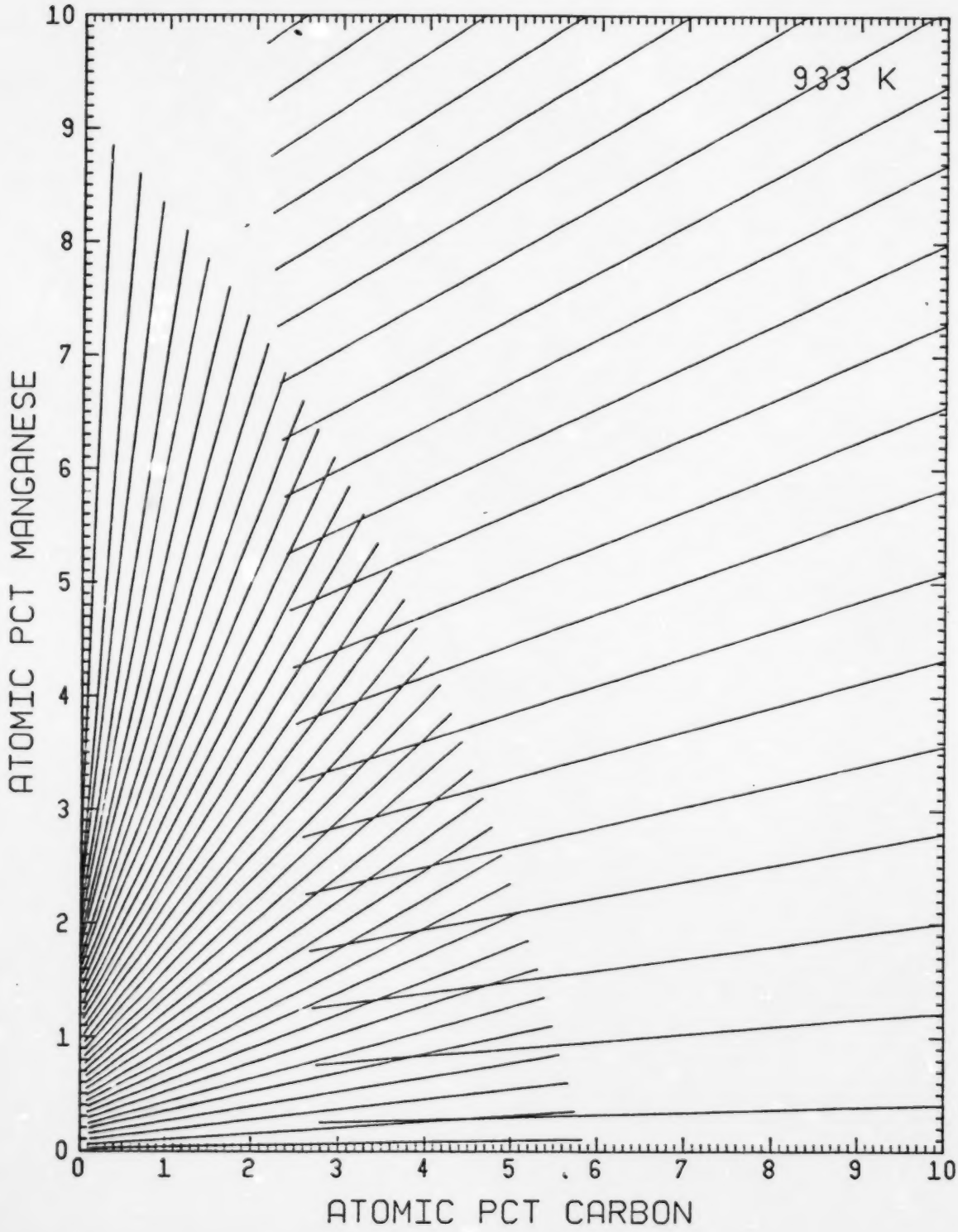
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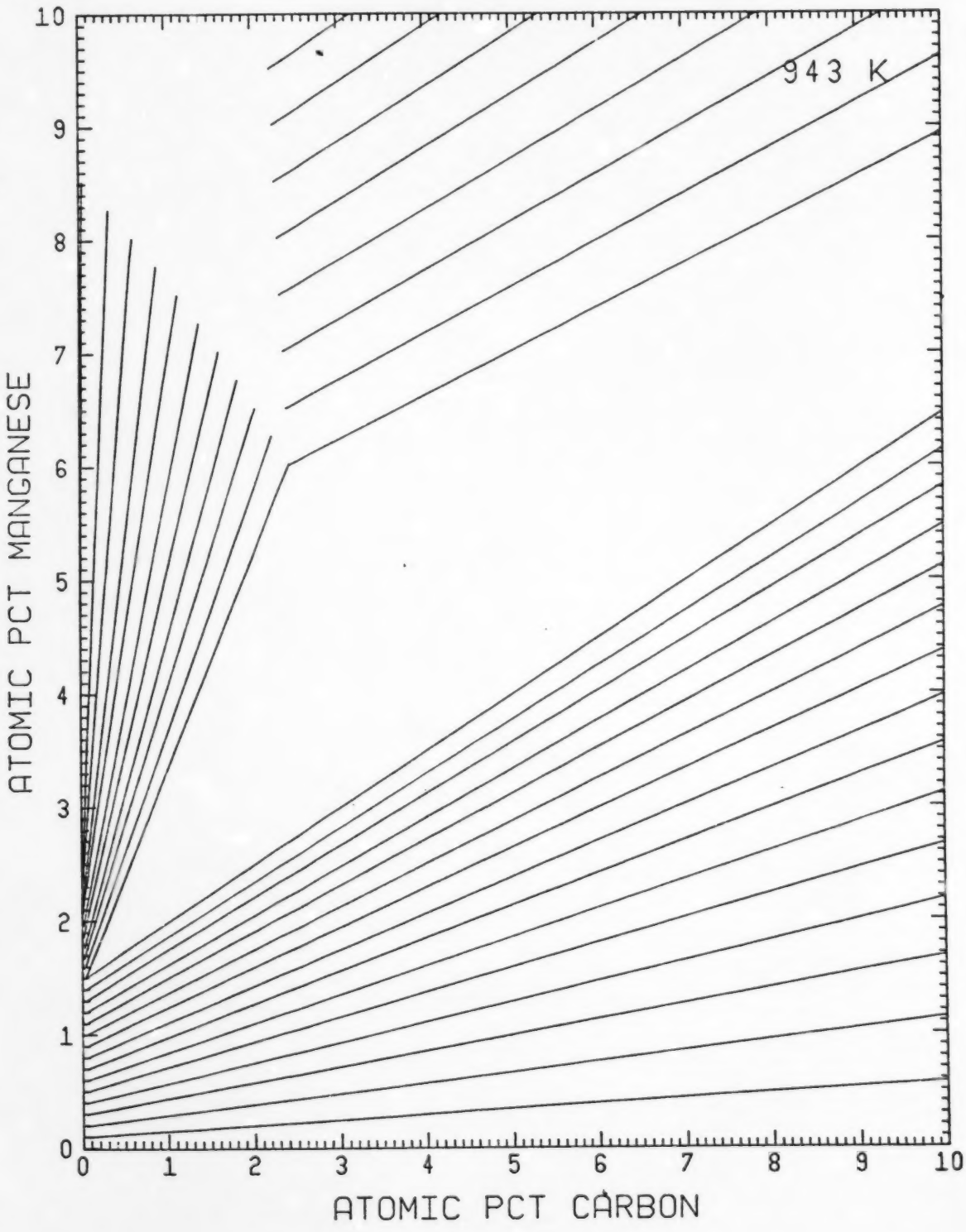
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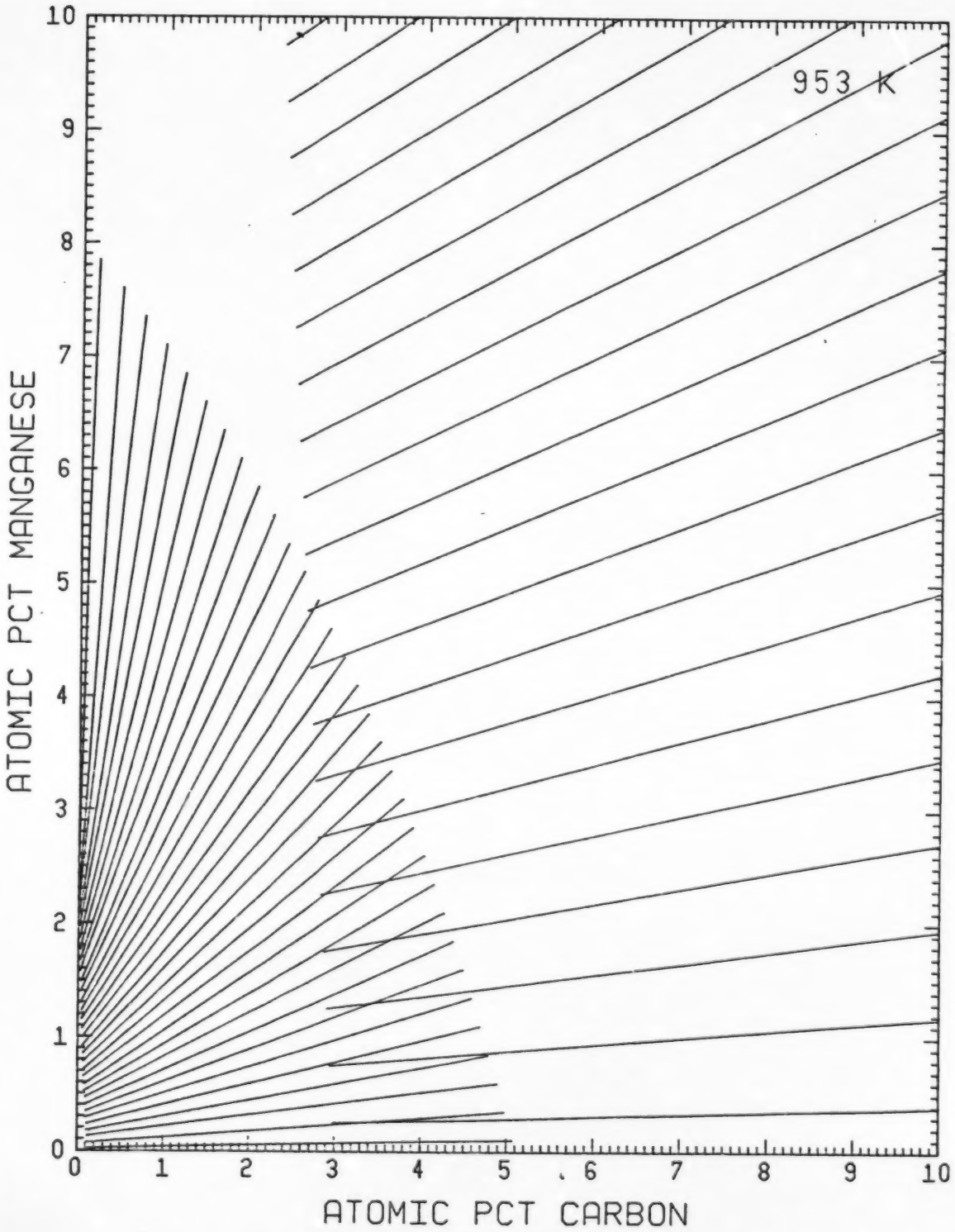
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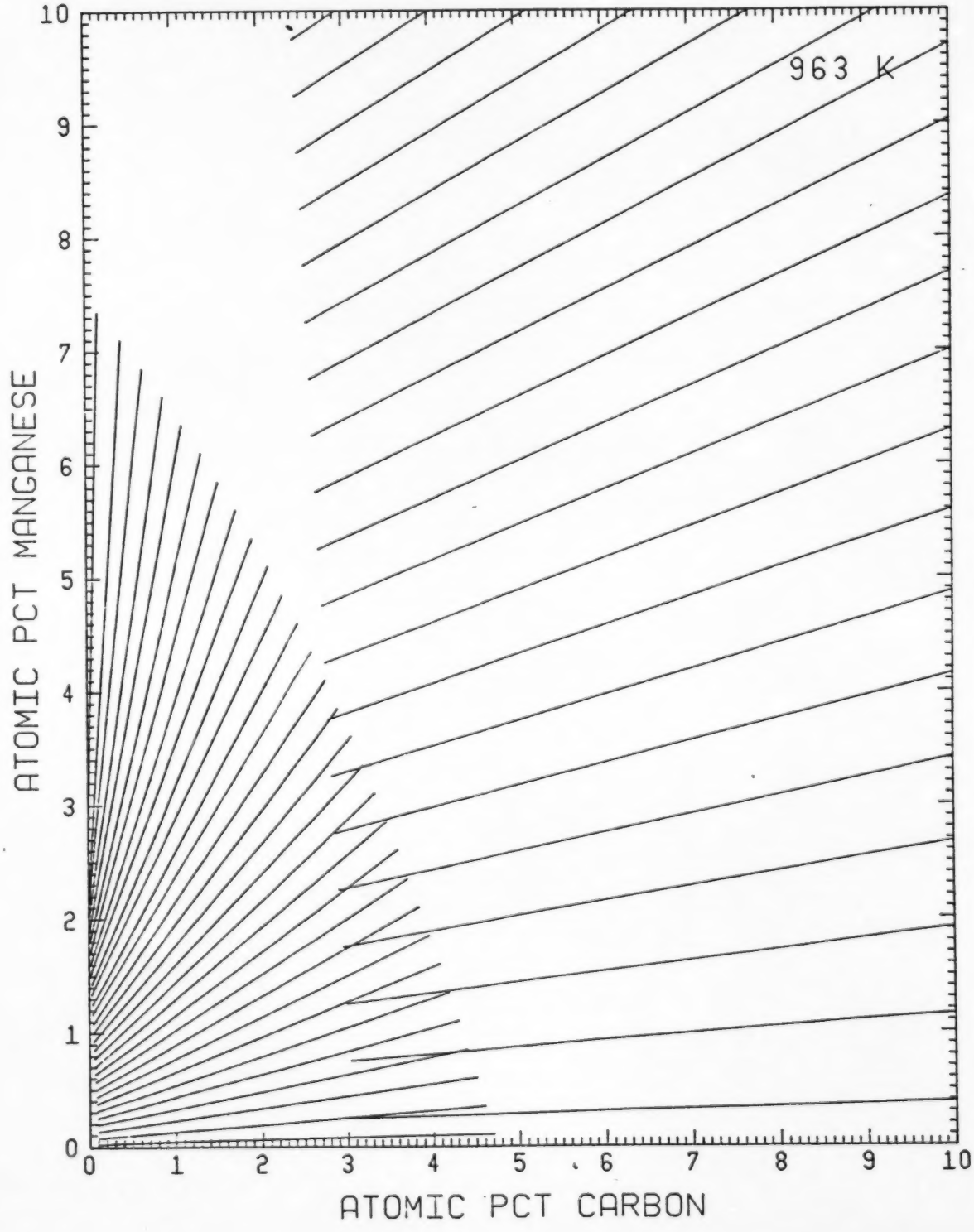
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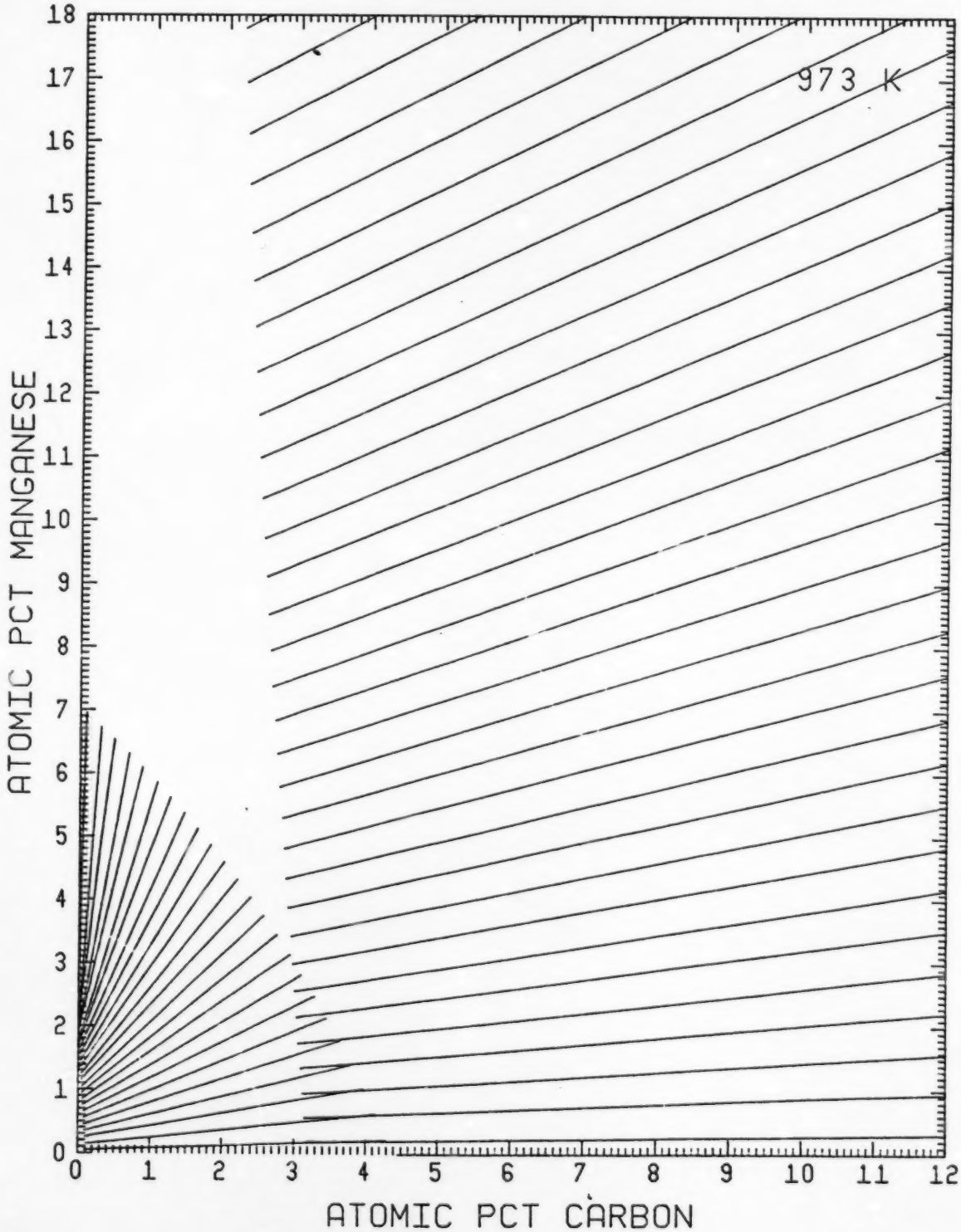
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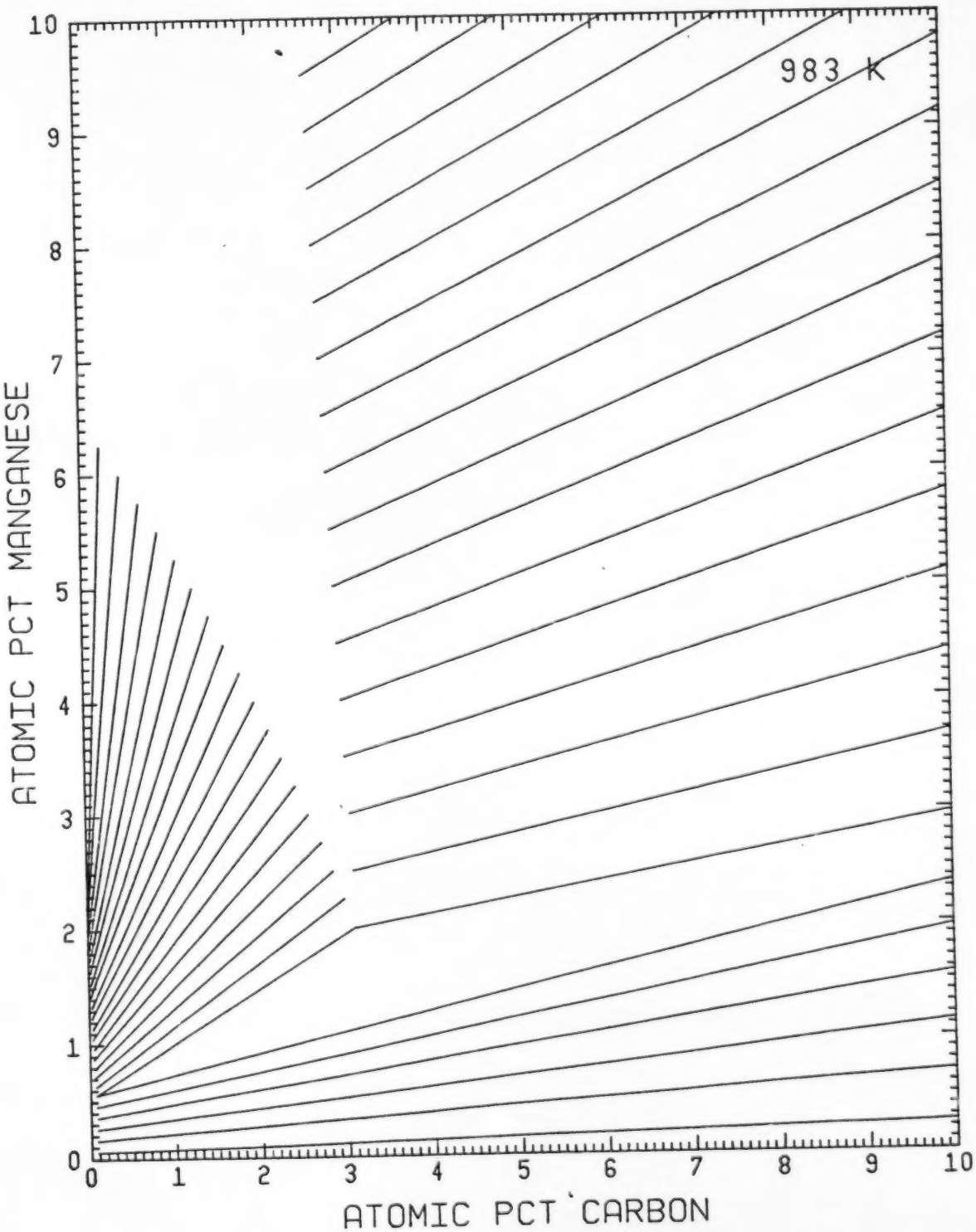
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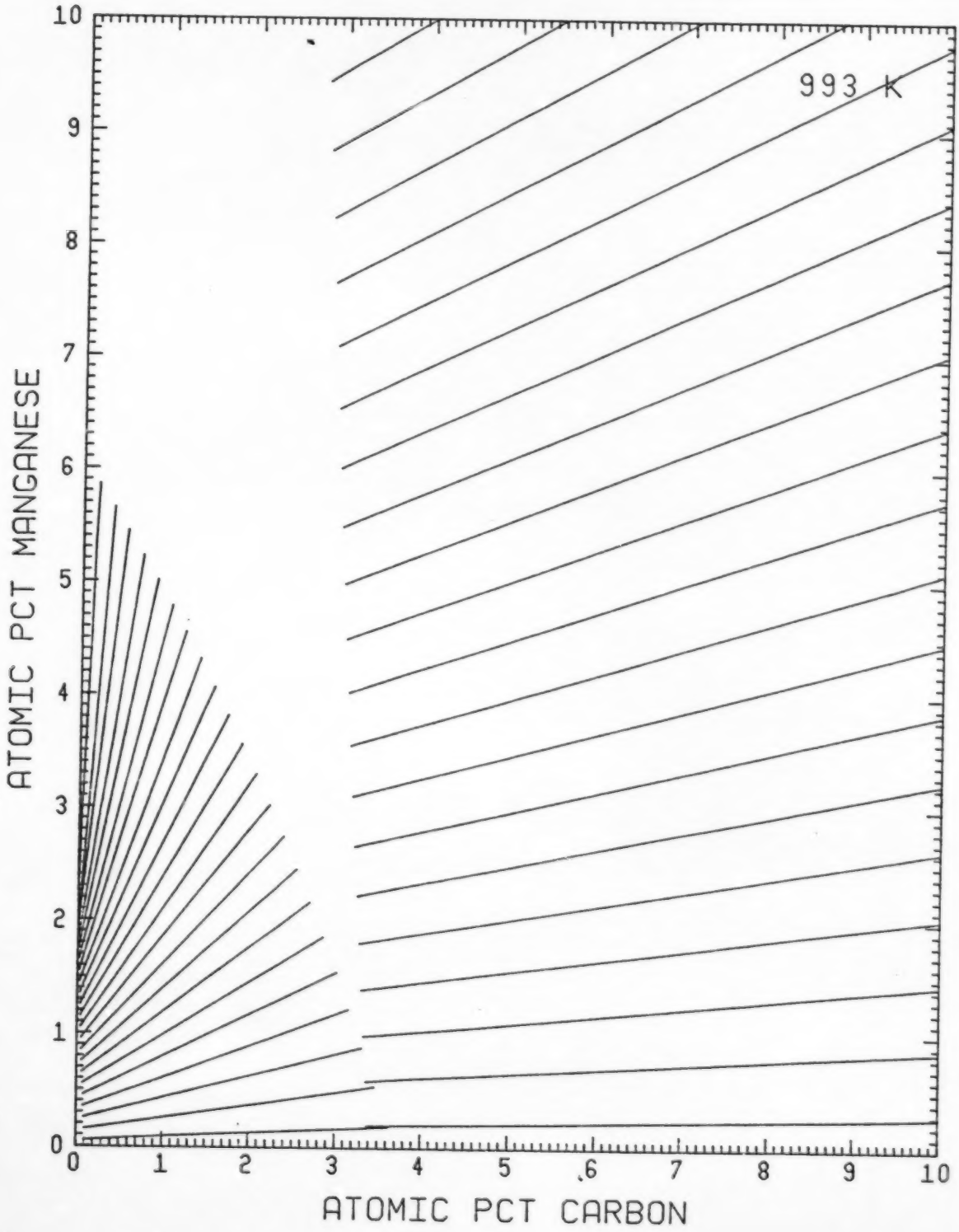
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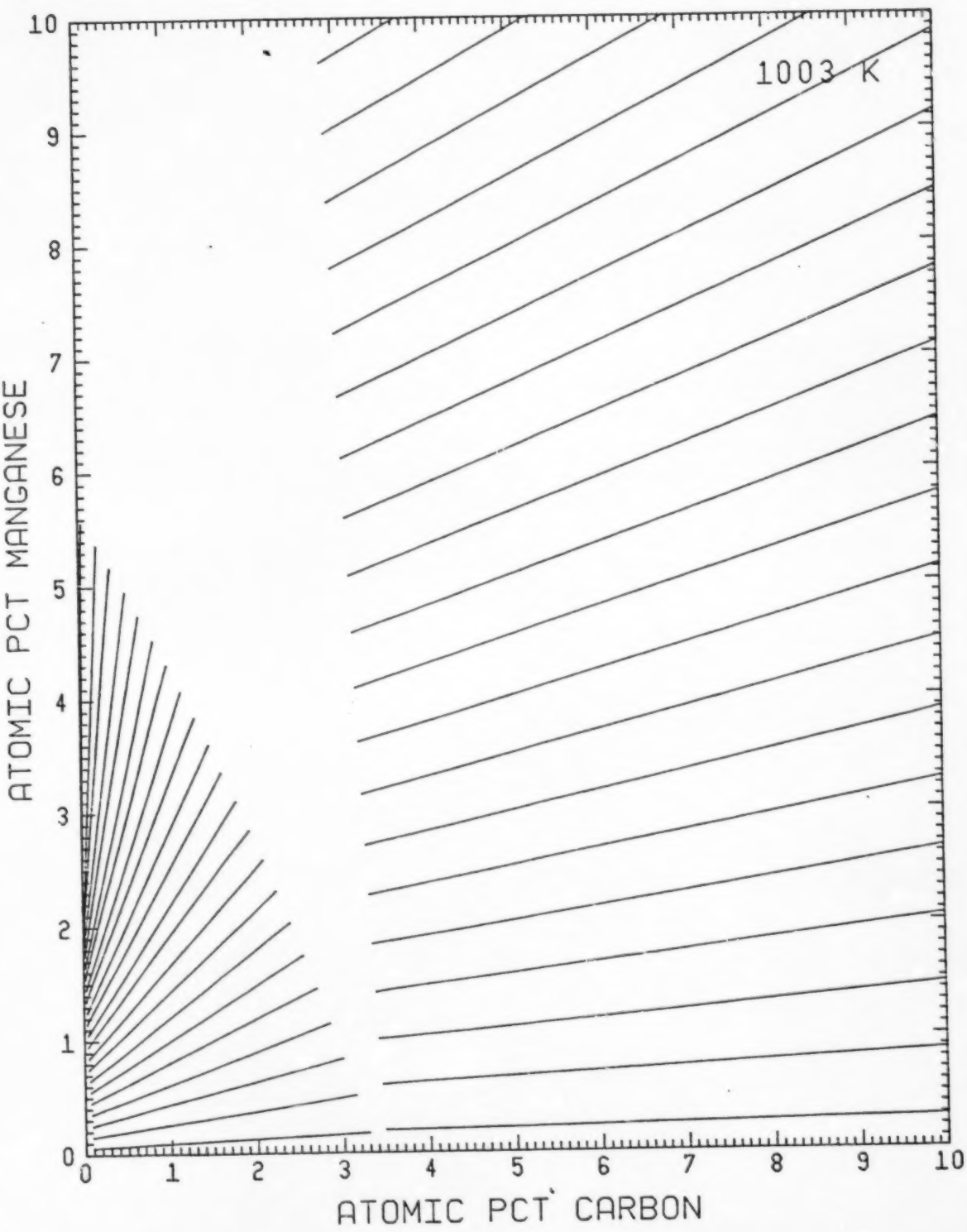
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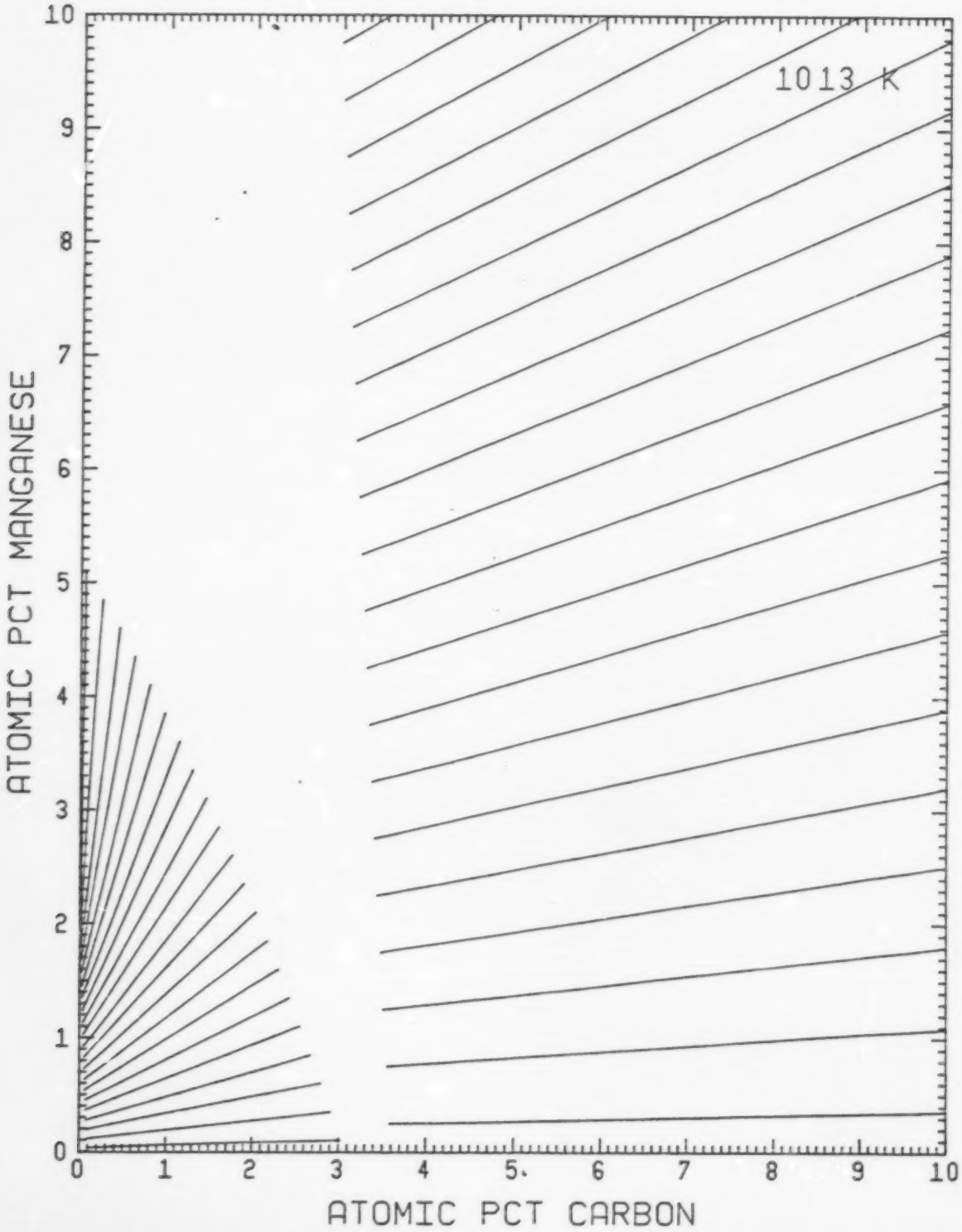
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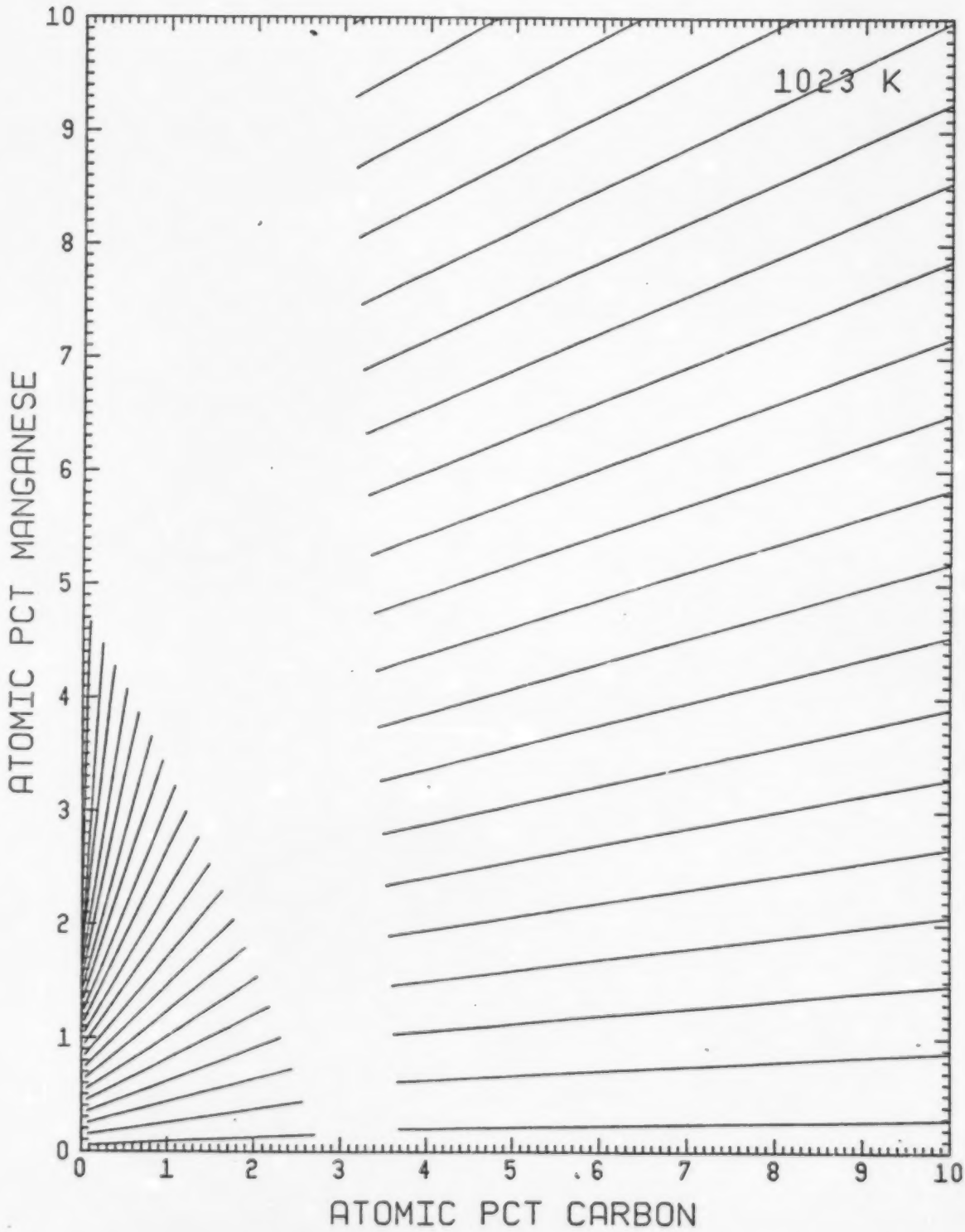
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